

## Large-Scale Production Results with Lead-Bearing Steels

*Since the first announcement of the development of free-machining open-hearth lead-bearing steel, interest in its application has grown rapidly, due largely to the fact that, in addition to greatly improved machinability, the chemical and physical properties of the particular type or grade of steel containing lead remain substantially the same as when untreated with lead. With the growing use of lead-bearing steels evidence of the improvement in production is being established and applied to the war effort. Some production results are given which compare outputs, using steels treated and untreated with lead.*

WE are gradually realising that the strength and efficiency of our fighting forces are largely governed by the extent and skill of the industrial organisation, translated into machines of war, available to them. We have ceased to estimate our strength on numbers alone and recognise that the factory, the organisation and the machine are much more important than at any previous time in the world's history. This war is proving that speed in action is of primary importance and that success depends largely upon the efficiency of the mechanised equipment at the disposal of the forces engaged rather than on the number of men on the battle front. To facilitate and maintain an adequate supply of the most modern implements of war the industrial power of the British Empire is being increasingly utilised and efforts are continually being made to speed up their production.

To speed up production new tools and equipment are desirable, but it would be absurd to wait until sufficient new machinery has been built to satisfy present needs, though progress in this direction has been remarkable. In general, however, most of the plants engaged on the production of implements of war must rely on machinery installed for peace-time activities, the majority of machines, therefore, are not new and their output is lower, other factors being equal, than the most modern machines. Under these circumstances it is necessary to determine ways and means of making the best possible use of the machinery installed in factories, and to continue the installation of new machines so that they can be employed in the most useful way as they become available.

In addition to making the best possible use of machine tools available, however, attention has been directed to materials in a further effort to speed up production, and considerable progress has been effected in this way by the development of materials which possess the required strength, but which can be machined at greater speeds without detriment to the material or the tools employed. Indeed, investigations have shown that in addition to greater output effected in this way, the life of cutting tools has been prolonged. This is especially true of lead-bearing steel, which is a development resulting from a new approach to the problem of improved machinability.

The importance of steel in supplying the means for defensive and offensive operations cannot be over emphasised, and efforts have been mainly directed to the improvement in machinability of this material with a view to increasing cutting speeds and thus raise the capacity of the machine tools available, whether new or old. Efforts to improve the machinability of steel have been made

over a long period, and as a result of much work the type of steel familiarly known as free-cutting steel was developed. The use of sulphur, phosphorus, silicon, nitrogen, and to some extent manganese, in steels to promote free machining characteristics were developed, and are being applied, particularly those containing sulphur and phosphorus as essential constituents. The results achieved with these free-cutting steels have been remarkable, but their field of usefulness is somewhat restricted as the principles upon which the free-cutting properties are based cannot be universally applied to the wide range of carbon and alloy steels used in war machines and equipment because of the adverse effect on toughness, corrosion resistance, forgeability or response to heat-treatment, either separately or in combination, according to the method adopted to improve machining.

A new approach to this problem resulted in the addition of a small percentage of lead to open-hearth steel, and evidence of the improved machinability of steel possessing a small lead content was first announced by Robbins.<sup>1</sup> Subsequently Nead, Sims and Harder<sup>2</sup> who were intimately associated with the development of this new free-cutting steel published authoritative data on this striking product. Tests have shown that the machinability of the majority of steels, to which suitable lead additions are made, is much improved, and a noteworthy feature of this development is that the chemical and physical properties of the particular steel remain substantially the same as when untreated with lead.

Although the amount of lead present is very small—0.20 to 0.25%—its influence on machining is most marked, and large scale production tests of a number of grades of steel, including free-cutting, mild and carbon steels and several alloy steels, all show decided improvement in machinability. The method of adding the lead is applicable to the manufacture of all steels made by melting and ingot casting methods. So small is the addition that an improvement in machinability can be noted when less than 0.10% lead is present, while an increasing effect is produced with the content increased to 0.3% or over.

Initially lead-bearing steel was used in considerable quantities as screw stock, but in a remarkably short time additions of lead have been made to grades of steel over a wide range with complete success, and in the steel industry lead is now regarded as an important alloying element for

<sup>1</sup> F. J. Robbins, "Character of Machine Performance of Lead-Bearing Steels," *Iron Age*, Vol. 142, pp. 28-33.

<sup>2</sup> H. Nead, C. E. Sims and O. E. Harder, "Properties of Free-Machining Lead-Bearing Steels," *Metals and Alloys*, Vol. 10, pp. 68-73.

the improved machinability of strong, heat-treated steels. A considerable amount of evidence has accumulated from production tests employing a multiplicity of machining operations under varying conditions, in which lead-bearing steels have greatly increased output. There is much more evidence, however, in the possession of those who are using these steels in everyday operation, and in view of their possible effect in assisting to speed up production more generally, some of the results obtained in practice will be of particular interest, especially as the improved production often exceeds the laboratory estimated maximum and has reached 100% or more in specific cases.

It will be of value to give full details of the method of production and production time of six articles from data supplied by users of lead-bearing free-cutting steel operating on a large scale.

## EXAMPLE 1.

Bottom Bracket Cycle Spindle on 4-Spindle Gridley from  $\frac{1}{2}$  in. dia. material.

	Super Free-Cutting Steel	Ledloy Free-Cutting Steel
Spindle speed	500 r.p.m.	700 r.p.m. (max. of machine)
Time each	59 secs.	28 secs.
Coolant	Soluble oil	Soluble oil
Tool steel	18% Tungsten	18% Tungsten
Increased production	—	110%

## EXAMPLE 2.

$1\frac{1}{2}$  in. dia. Screwed Bush on Pitter Semi-Automatic.

	Super Free-Cutting Steel	Ledloy Free-Cutting Steel
Spindle speed	400 r.p.m.	600 r.p.m. (max. of machine)
Time each	2 mins. 50 secs.	2 mins. 20 secs.
Depth of cut	$\frac{1}{8}$ in. $\times$ .0035 in. feed per rev.	$\frac{1}{8}$ in. $\times$ .007 in. feed per rev.
Increased production	—	17% exclusive of tool life effect

On trying to turn the super free-cutting steel at the faster speed the tool burnt up on the first one tried, but Ledloy continued to run without trouble.

## EXAMPLE 3.

Grease Nipple on  $\frac{1}{2}$  in. Single-Spindle B.S.A. from .340 in. a/f material.

	Super Free-Cutting Steel	Ledloy Free-Cutting Steel
Spindle speed	2,495 r.p.m.	4,140 r.p.m.
Spindle speed (screwing)	875 r.p.m.	1,580 r.p.m.
Time each	20 secs.	13 secs.
No. of parts per hour	180	276
Increased production	—	Over 53%
Increased tool life	—	300%

## EXAMPLE 4.

Whitworth Set Screw  $\frac{1}{8}$  in. dia.  $\times$   $2\frac{1}{2}$  in. long on Single-Spindle Cleveland from .601 in. a/f material.

	American Screw Stock	Ledloy Free-Cutting Steel
Spindle speed turning	490 r.p.m.	693 r.p.m.
" " screwing	152 r.p.m.	200 r.p.m.
Time each	95 secs.	69 secs.
Tool life Chasers, Herbert die head	2 $\frac{1}{2}$ days	4 days
Increased production	—	37-6%

## EXAMPLE 5.

Bolt  $\frac{1}{2}$  in. dia.  $\times$  2 in. long on 4-Spindle Acme Gridley from .601 in. a/f material.

	American Screw Stock	Ledloy Free-Cutting Steel
Spindle speed turning	896 r.p.m.	1,126 r.p.m.
" " screwing	296 r.p.m.	373 r.p.m.
Time each	21 secs.	13 secs.
Production per hour	171	277
Increased production	—	62%
Tool life	—	300%

## EX. M.F.L.E. 6.

$\frac{1}{4}$  in. R.S.F. Double Chamfer Locknut Blanks on R.D. Size 0-0 Machine (Single Spindle) from .448 a/f material.

	3 S.I.	Ledloy 3 S.I.
Spindle speed	2,140 r.p.m.	2,140 r.p.m.
Time each	18 secs.	10 secs.
Increased production	—	44%
Tool life	—	50%
Tool life on 2nd operation (tapping)	—	Increased 200%
Spindle speed tapping	950 r.p.m.	2,200 r.p.m.

In assessing the value of the lead-bearing steel Ledloy, from the above data, it must be remembered that the condition of the machine or machines which were used

for the production of the parts in question varied from shop to shop, some of the machines being relatively old, while others were of the newest type. Reports indicate that the better the machine, the better the improvement in production which results from the use of a lead-bearing steel. This, of course, is to be expected because rigidity has a considerable effect on production capacity.

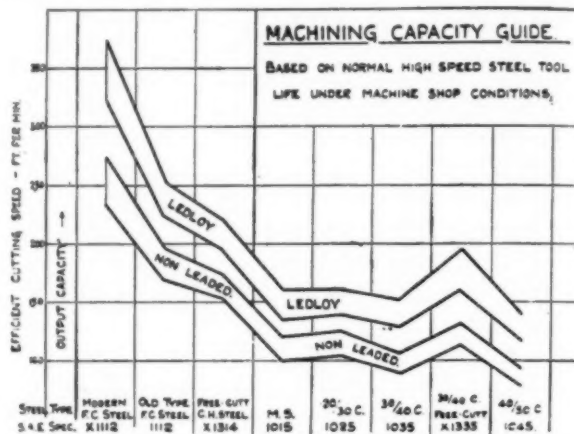


Fig. 1.—Economic cutting speeds for different grades of steel, with and without a lead constituent.

Numerous additional examples of increased production could be cited from the use of lead-bearing steel. In each case it is necessary to take advantage of the increased cutting speed that lead-bearing steel permits and, in order to provide a reasonable guide, the chart, Fig. 1, gives for each class of steel what accumulated experience from large scale production has shown to be the most economic cutting speed for average work, which will permit of reasonable tool life and, in general, optimum efficiency. The chart shows the relation between cutting speed and analysis of the steel both for leaded and non-leaded grades. It will be appreciated, however, that since the condition of the machine tool, the type of finish required on the component, the accuracy of its dimensions, and the type of machining operation carried out, all affect the cutting speed, reasonable allowance should be made for these variables.

The examples given are concerned with work associated with export activities, but it will be appreciated that the saving of machining time possible in connection with war work opens an important field of usefulness for lead-bearing steels. The needs of the Forces in mechanical equipment, guns and shells, lend themselves to production on a quantity basis, and it is not surprising that both long- and short-range tests have been carried out with a view to the application of these steels. At first small batches of lead-bearing steel bar were put under trial at various works. Preliminary results were obtained for small calibre shells, and the following figures are given from a 16-day record of tool regrinds and production figures for forming and parting a batch of lead-bearing steel bars on a Gridley machine in the manufacture of 40 mm. shells. A comparison is made with a batch of steel possessing average machining qualities of normal deliveries of shell steel to the appropriate specification. The results obtained were:—

Material.	Cutting Speed.	Feed.	Time.
Lead-bearing steel	330 r.p.m.	.0065 in.	2 mins.
Lead-free steel	210 r.p.m.	.0065 in.	2 mins. 50 secs.

Life of the form tool, taper tool and part-off tool was increased by 30%. Normally, in the production of these shells, the drills are reground each day, but in this case it is only normal drill sharpening which is required, whereas, when machining lead-free steel, the drill is burned and

requires a certain amount of length ground off before it can be used again. Thus, including time lost setting and grinding tool, the average production from lead-bearing steel was 25 per hour, and from lead-free steel 18.3 per hour, showing an improvement in production of over 30%.

Other operations also showed an increase in the life of tool and in production, and the following data gives a comparison of old times and feeds compared with new times and feeds when machining lead-bearing steel.

Machine	Lead-free Steel		Lead-bearing Steel		Increase Production
	Cutting Speed	Time	Cutting Speed	Time	
Capstan ..	539 r.p.m.	2 mins. 15 secs.	736 r.p.m.	1 min. 40 secs.	35%
Capstan ..	301 r.p.m.	2 mins. 15 secs.	535 r.p.m.	1 min. 30 secs.	50%
Tapping ..	92 r.p.m.	35 secs.	135 r.p.m.	25 secs.	40%
Tapping ..	185 r.p.m.	30 secs.	260 r.p.m.	25 secs.	20%

At another works difficulties were occasionally experienced in machining the normal shell steel in the production of 40 mm. shells. A 5-ton lot of lead-bearing steel was put on trial. In the light of this firm's previous experience this trial lot was commenced at the same cutting speed used on lead-free steel. No trouble was experienced, and the finish on the shell bodies was superior to that on bodies selected at random from normal production. The conditions, however, were not such as to provide a suitably severe test of the lead-bearing steel, and it was suggested that the machine on which the trial was being made should be run at its maximum speed. After some objection by representatives of the firm who were afraid that severe damage to the tools would result, this was done, with the result that the machining time was reduced from 2 mins. 35 secs. to 1 min. 44 secs., i.e., an increase in production of 33%. Tool life was satisfactory. The finish on the shells produced at this increased speed was excellent, and on issuing from the machine after this final operation they were much cooler than those coming from a machine running more slowly with normal shell steel.

In view of the success achieved at the above works it was considered advisable to proceed with a large scale production test, and for this purpose the trial was extended to cover the use of some 50 tons of lead-bearing steel. Comparison was made with normal lead-free steel, of average machining properties, supplied to the required specification for 40 mm. shells. Complete data were prepared covering a prolonged test which included the life of each tool used in the various positions, as well as the increasing of production due to the use of lead-bearing steel. During the period of the test the production time of each shell was 1 min. 45 secs. for lead-bearing steel against 2 mins. 26 secs. for lead-free steel, and, taking into account the changing of tools for the greatly increased number of regrinds, the average production per hour was 19.5 for lead-bearing steels against 9.5 for lead-free steels. Four machines were used in this trial, and basing the machine efficiency on a maximum of 80% the results obtained indicated a machine efficiency of 72.75% when using lead-bearing steel against 48.2% for lead-free steel. Under the conditions of this trial the data show that it is possible to obtain an increase in production of 100%, together with an increase of tool life of approximately 200% (the number of tool changes using lead-bearing steel was about one-third that necessary when running on lead-free steel of similar composition).

In addition to low carbon steels many tests have been carried out on a variety of types containing lead with a view to obtaining information on the effect of this constituent on mechanical properties, machinability, etc. An interesting investigation of this character was carried out on over 20 types of lead-bearing alloy steel. The method employed was to make half of the cast lead-bearing and the remaining half non-lead-bearing. In every case steels containing lead showed considerable improvement in

machinability of high commercial value. The results obtained from three types of alloy steel in common use are given in the following table, and can be regarded as representative of the whole series tested in this investigation.

ANALYSIS OF SAMPLES.

Sample	C.	Mn.	Si.	S.	P.	Ni.	Cr.	Mo.	Pb.
1	0.18	0.64	0.26	0.027	0.014	3.45	—	—	0.01
1a	0.17	0.66	0.28	0.027	0.014	3.29	—	—	0.17
2	0.46	0.92	0.28	0.028	0.012	3.47	—	—	0.02
2a	0.48	0.93	0.30	0.025	0.015	3.45	—	—	0.20
3	0.41	0.74	0.22	0.023	0.013	1.73	0.65	0.33	0.02
3a	0.42	0.75	0.21	0.022	0.012	1.74	0.65	0.36	0.17

a—The Lead-bearing Samples.

MECHANICAL PROPERTIES.

Sample	Treatment	Tens. Strength Tons/sq.in.	Yield Pt. Tons/sq.in.	Elongation %	Red. Area %	Brinell No.	Charpy Impact (ft. lbs.)
1	Normalised	38.8	26.0	31.0	63.6	163	44
1a	"	39.3	28.0	31.5	61.0	163	39
2	Annealed	52.0	31.8	22.5	46.4	217	14
2a	"	52.6	33.2	23.0	47.5	217	14
2	Heat-treated O.Q. & Temp.	56.5	38.1	20.6	42.5	226	24.5
2a	"	55.2	41.5	20.0	45.5	223	23.5
3	Annealed	61.0	—	17.0	47.5	241	13.5
3a	"	63.2	—	17.5	46.5	241	14.0
3	Heat-treated	67.5	59.2	18.0	56.5	270	37.5
3a	"	63.5	53.0	18.0	53.0	255	33.5

MACHINABILITY DATA.

Sample	Normalised or Annealed			Oil Quenched and Tempered		
	Sawing Index	Drilling Index	Combined Improvement	Sawing Index	Drilling Index	Combined Improvement
1	0.894	1.03	—	—	—	—
1a	0.626	0.679	47%	—	—	—
2	0.24	1.18	—	1.09	1.355	—
2a	0.600	0.811	42%	0.776	1.125	28%
3	0.815	1.261	—	0.948	1.29	—
3a	0.665	0.855	37%	0.633	1.22	21%

Among the many further tests to which lead-bearing steels have been submitted is a series concerned with corrosion. A comparison has been made between lead-bearing free-cutting steel, lead-bearing mild steel and lead-free mild steel; specimens of each type being exposed for one month to the outdoor atmosphere, salt spray and damp storage. The lead-bearing free-cutting steel, compared with lead-bearing mild steel had a similar lead content but a much higher sulphur and phosphorus content, and the results of the tests show that it was slightly more susceptible to corrosion. The lead-free mild steel of moderately high sulphur and phosphorus content corroded a little more than the lead-bearing free-cutting steel on outdoor exposure, but a little less under damp storage conditions. It was concluded that the difference in corrosion resistance between lead-bearing and ordinary steel is small and of no significance.

The data and results given in this article have been selected from the mass of evidence accumulated during the past two years, but much more is in the possession of those who are using lead-bearing steels in everyday operation. These steels are no longer only of interest metallurgically; in this time of emergency they have become of increasing interest to the engineer, and in many machine shops to-day lead-treated steels have become the normal material for the production of a remarkably wide range of components connected with the war effort. Their improved machinability in comparison with similar lead-free steels enables greatly increased production, increases the efficiency of machine tools, and is contributing substantially to the speeding up of supplies of ammunition and mechanised equipment to the Forces.



## A Simple Gas Testing Appliance

By J. G. LEWIS, M.A.

**C**OMMON methods of setting gas proportioning arrangements used for making protective furnace atmospheres are to make adjustments with reference to results of gas analysis or to the finish on a sample of the metal annealed in a miniature furnace containing the protective atmosphere. Both methods are too slow for convenience in the works and competent Orsat users are getting scarce. The composition of the gas atmosphere can usually be controlled by checking the hydrogen content only, using a katharometer, but instruments of this kind are expensive and are now almost unprocurable. The instrument described below is quick in action, cheap, and simple enough for any furnace operator to use. It functions by diffusion and is operated principally by the hydrogen present in the gas. Although it does not analyse the gas, it enables the composition to be maintained sufficiently constant for certain bright annealing processes.

The apparatus is a form of diffusion apparatus frequently used to demonstrate the diffusion of gases to schoolboys, and is made up as shown in the sketch. The porous pot is a close grained unglazed pot as used in wet batteries.

The instrument is manipulated as follows: Place the beaker over the porous pot and turn on the gas from the atmosphere plant providing the protective atmosphere, preferably at a constant and fairly rapid rate. The pressure shown by the manometer rises to a maximum, and then gradually falls back if hydrogen is present in the gas. The maximum pressure, reached in 20-30 secs., is a function of the gas composition, and should now be read. Then turn off the gas and remove the beaker. The manometer settles down to zero again in about 4 mins. when another reading can be taken if required. The pressure swings, which indicate the limits of useful gas composition, are soon found by trial and error or by calibration against gas analysis apparatus.

The theory of the instrument is based on Graham's law of diffusion, namely that the rate of diffusion of a gas through a porous pot is proportional to the partial pressure difference of that gas across the pot, and inversely proportional to the square root of the gas density. Taking 100th of an atmosphere as the unit of pressure and the density of hydrogen as one and neglecting moisture differences, the state of affairs at the start of a test is shown in Tables I and II for the two gas compositions as given below:—

Composition of Gas given in Table I: Hydrogen, 1.5%; Carbon Monoxide, 0.5%; Carbon Dioxide, 13.5%; Nitrogen, 84.5%.

Composition of Gas given in Table II: Hydrogen, 3.0%; Carbon Monoxide, 1.0%; Carbon Dioxide, 13.0%; Nitrogen, 83.0%.

It will be fairly obvious when glancing through these tables, in particular with reference to the figures given under the 5th column in each table, that it is the percentage of hydrogen which most greatly affects the inward rate of diffusion, as this is proportional to the partial pressure, or the percentage of hydrogen in the gas divided by the square root of the density—which for hydrogen is unity. Hence the higher the percentage of hydrogen in

the gas, the greater the total inward diffusion and the greater the difference shown on the water or oil manometer.

TABLE I.

Gas	Gas Density	Partial Pressure in Burned Gas	Partial Pressure in Air (inside porous pot at start of test)	Inward Diffusion rate proportional to	Outward Diffusion rate proportional to
Hydrogen .. ..	1	1.5	0	1.5 $\sqrt{1} = 1.5$	—
Nitrogen .. ..	14	84.5	79	5.5 $\sqrt{14} = 1.47$	—
Carbon Monoxide ..	14	0.5	0	0.5 $\sqrt{14} = 0.13$	—
Carbon Dioxide ..	22	13.5	0	13.5 $\sqrt{22} = 2.88$	—
Oxygen .. ..	16	0	21	—	21 $\sqrt{16} = 5.25$
Total .. ..	—	—	—	5.98	5.25
Difference (Inwards) ..	—	—	—	0.73	—

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TABLE II.

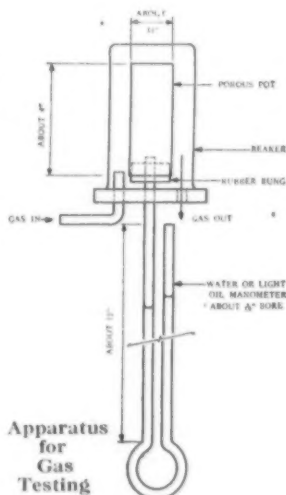
Gas	Gas Density	Partial Pressure in Burned Gas	Partial Pressure in Air (inside porous pot at start of test)	Inward Diffusion rate proportional to	Outward Diffusion rate proportional to
Hydrogen .. ..	1	3	—	3 $\sqrt{1} = 3$	—
Nitrogen .. ..	14	83	79	4 $\sqrt{14} = 1.07$	—
Carbon Monoxide ..	14	1	—	1 $\sqrt{14} = 0.27$	—
Carbon Dioxide ..	22	13	—	13 $\sqrt{22} = 2.77$	—
Oxygen .. ..	16	—	21	—	21 $\sqrt{16} = 5.25$
Total .. ..	—	—	—	7.11	5.25
Difference (Inwards) ..	—	—	—	1.86	—

An example of the use of this instrument on a large scale commercial plant for the clean annealing of copper illustrates the order of the manometer readings obtained. On one furnace the gas is found to be satisfactory when the maximum positive pressure shown by the manometer is between 1 and 4 ins. water gauge. With 5 ins. water gauge maximum pressure the gas is just inflammable. Troubles due to gas of the wrong composition have practically stopped since the instrument was put into service and the time required to reset valves after an overhaul has been reduced to half an hour or less.

### Furnaces for Reheating and Heat-treatment of Aluminium and its Alloys

FURNACES for reheating and heat-treatment of aluminium alloys demand the consideration of features which are peculiar to these alloys. For practical purposes the temperature range of these furnaces is from 100° to 600° C. Artificial ageing is carried out at temperatures between 100° and 200° C., whilst the solution treatment involves temperatures ranging from 400° to 560° C. There is a critical temperature for most heat-treatments of these light alloys which varies with their composition, and consequently precise temperature control is essential. Also, the duration of a given heat-treatment often varies with the form of the material to be treated, and is, in some cases, longer when the material is in the form of a casting than when it has been wrought.

The design, construction and operation of furnaces for the annealing, reheating, solution treatment and artificial ageing of these alloys are surveyed in a recent issue of "Aluminium Technique," in which are discussed the types of furnaces developed for aluminium alloys, the heating media used, the use of salt baths, and design and construction details. The brochure contains much useful information of a practical character and some excellent illustrations of typical furnace installations. Copies are available from Aluminium Union Limited.





# METALLURGIA

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INCORPORATING "THE METALLURGICAL ENGINEER"

## Production for the War Machine

**T**HE industrial organisation of this country has been severely criticised from time to time for its apparent inability to adjust itself readily to the conditions imposed upon it by war. Much of the criticism has been directed against the Ministry of Supply, but in effect, any shortcomings which it implies reflects upon the engineering industry as a whole. It is doubtful, however, whether many who are so ready to criticise the efforts already made really appreciate the problems encountered in transforming the country into an arsenal. It must be remembered that, at the opening of hostilities, the whole industry of Germany was mobilised and harnessed to the Nazi war machine. This machine had been developed and constructed over many years during which German genius had concentrated on the development of fighting technique likely to shatter the most stubborn resistance.

Germany organised her forces for rapid movement and equipped them to take every advantage of the new technique she had developed. Years of training produced the industrialised soldier from which emerged forces mechanised in mobility, in fire-power and scientifically equipped. Faced with this war machine, France collapsed, Belgium and Holland surrendered, Norway and Yugoslavia abandoned organised resistance, and Poland became enslaved. In comparison with this forward movement of a highly mechanised army, Britain fought a succession of magnificent retreats and masterly evacuations. This was not due to inferior courage or morale but to the fact that the British forces as well as the forces in the conquered territories were not adequately equipped to meet the new war technique evolved by Germany. It was this lack of equipment which caused the severe criticism of the Ministry of Supply, but whatever justification the critics may have had, it seems a pity that they did not make themselves sufficiently heard long before the outbreak of war.

To harness the industry of a country to its fighting forces is not an action that can take place by the stroke of a pen; Germany was building up and perfecting her schemes over a period of seven years before she attacked Poland, and, when Britain declared war, relatively little had been done to organise and mobilise industry to combat the resources of mechanical power then in the hands of Germany. It must be apparent to any normal person that to transform industry from a peace to a war footing, when war has been declared, is much more difficult because a condition of emergency then exists in which time becomes a vital factor. Industry must then be organised to supply immediate needs rather than to meet the requirements of a comprehensive scheme to equip the forces in the light of modern mechanised technique. It is not surprising, therefore, that difficulties have been encountered in obtaining supplies.

Fortunately, there was one place in which Britain effectively defeated the German force; in the Battle of Britain our machines proved as supreme as our pilots. This battle emphasised the need for more and more aeroplanes, not only to maintain an adequate defence, but to build up something of a parity with German air force, to ensure an effective offensive. Much has now been done, and, with the aid of the United States aircraft

industry, production for this branch of the forces has been steadily mounting, and the results are seen in the increasing offensive over territory occupied by the enemy as well as against enemy sea transport. For some months now the Royal Air Force has delivered almost unbroken air-raids on Germany by night and occupied France by day, which shows an encouraging increase in Britain's ability to hit back. The reserves of aeroplanes, airmen and ammunition required to maintain this offensive are very great, and it is encouraging to realise that the strides made in increasing output have contributed largely to this end.

Production for the Air Force, however, is only a part of the scheme of production for the war machine. It is probably true that in the beginning the Allies envisaged the conduct of this war with the weapons of the last war. But Germany had decided to stop trench warfare by carefully placed artillery, both light and heavy tanks, the dive-bomber and the development of parachute troops. The weight of machines and fire-power rendered trenches untenable; it was necessary, therefore, not only to organise industry to supply the means to combat this power of specialists, but to reorganise the fighting forces, so that each fighting unit would have equipment and specialists comparable with the enemy.

It is not necessary to discuss what has now become recognised as essential to field forces, such as the many and varied sizes and types of tanks; heavy, medium and light artillery; anti-aircraft guns; anti-tank guns; machine and sub-machine guns; flame and grenade throwers; light and heavy mortars; as well as many other forms of defensive and offensive weapons, including the rifle and automatic pistol. In addition, the transporting facilities for equipment, personnel and services must be ample for any emergency.

Adjusting the industrial organisation of a country to meet the demands for these forces is no light task. It is true that these units lend themselves to production on a quantity basis, but much initial work of a highly skilled nature must be done before mass production can commence. Each particular component becomes a vital unit in the maintenance of a production schedule, and any failure due to material or the human element will impede progress. And it is too much to hope that this does not occur. However efficient a works department may be, if the personnel comprises a dozen or more, a percentage can be regarded more or less as passengers. Even among executives some are square pegs in round holes, and these tend to impede progress and thus interfere with output. On the other hand, the department is dependant upon ample supplies of material possessing the required characteristics for producing the component and failure to obtain delivery, or the delivery of an unsuitable grade, impose restrictions on output which have a direct effect on the production of the major unit.

Big demands are being made on industry to supply the needs of the Navy; despite the losses suffered by this branch of the Services, this part of the machine, in fighting power, now far exceeds that at any other period in British naval history. The never-ceasing struggle to bring materials of war and food supplies to these shores, against the persistent efforts of U-boat and long range bomber, has resulted in constant and inevitable losses. Industry is making valiant efforts to make good these losses and

new ships are now being built at greatly increased speed; just as important as the construction of new vessels is the maintenance of existing vessels in a seaworthy condition, here again the industrial organisation of this country has developed enormously.

In meeting the increasing demands for greater production the industrial organisations of Britain are being ably supported by the industry of the whole British Empire and, by means of research and experiment, efforts are continually being made by improved designs and materials to increase production results. While there is

still room for criticism, the results enable us to derive satisfaction from what has been done, but a still greater effort is needed to increase production. Only when we shall have matched machine against machine, eclipsing the German machine with the British, as the Spitfire has eclipsed the Messerschmitt, shall we achieve the victory to which we are dedicated. Let us therefore harness our efforts more effectively to the great war machine and ensure that, whether in the factory, shipbuilding yard or office we are making a maximum contribution to its success.

## Die Casting Speeds up Aircraft Production

THE problems associated with aircraft production during this war are in some respects no different from those experienced during the last war, when production of aircraft was speeded up. The quantities required then, however, were not near so great as now, and the difference in manufacturing methods is largely one of volume and a great difference in design. To-day, designs lend themselves more to quantity production, and the many years' experience in building and testing aircraft has tended towards the elimination of complicated components. Important highly stressed parts, for instance, the manufacture of which had previously necessitated lengthy and expensive "built-up" constructions, are now being produced virtually in single operations as aluminium alloy gravity die castings.

An example, illustrated, is a control member—a component in a modern high-performance aircraft—which is subjected to rigorous conditions in service for which only a high-strength alloy is suitable. The material used, NA226 alloy, has been specially developed for such purposes by the Northern Aluminium Company, Ltd. It is a casting alloy of high strength and guaranteed high elongation, and its value is extended by virtue of its suitability for making straightforward castings by the gravity die-casting process. The alloy is produced in conformity with three DTD specifications—361, 304 and 298, the production of the component referred to being governed by DTD304.

Details of minimum requirements regarding mechanical properties laid down in this specification are given in Table 1, together with the heat-treatment to which the material is subjected.

TABLE 1.  
TENSILE TEST FIGURES AND HEAT-TREATMENT LAID DOWN IN AIR  
MINISTRY SPECIFICATION DTD. 304.

Condition of Alloy.	0.1% Proof Stress.	Ultimate Tensile Stress.	Elongation % on 2 in.	Heat-treatment.
	Tons/Sq. In.	Tons/Sq. In.		
Sand cast ..	14.0*	18.0	4	Heat at 525°-545° C. for 12-16 hours. Quench in water. Reheat at 120°-170° C. for 12-24 hours. Cool in air.
Chill cast ..	17.0*	24.0*	9*	

\* These figures are given in specification for information purposes only.

The component, which measures 9½ in. × 5½ in., is illustrated in Fig. 1. It is cast in gravity dies, the only machining necessary being that required to size the hole in the base. It is thus an admirable example of the

simplicity of die-casting methods of production as compared with built-up constructions involving, possibly, flanged and formed sheet, forgings and tube, with the attendant machining, drilling, riveting and welding. Drawing-office work and scheduling are also reduced by the elimination of a number of component parts. The shock-resisting properties of the final product are revealed by the following data.

In submitting a production casting to the destruction test indicated in Fig. 2, in which a 7-lb. hammer was used, the test was extended until one top lug was fractured completely as shown, and the second lug bent until cracking first appeared. Calculations made on the assumption of ideal bending conditions, i.e., no strain at the neutral axis, and based on measurements of the bend radius and thick-



Fig. 1.—The gravity die-cast component.



Fig. 2.—The same component tested to destruction.

ness of the casting at the point where fracture commenced, gave an elongation figure of 15% over the length involved in the bend. These results were confirmed by tensile tests, the elongation values averaging 13%, 14.5% and 17% over gauge lengths of 2 in., 1 in. and ½ in., respectively.

The figures reveal the eminent suitability of the material for members likely to be subjected to severe stresses in service, and thoroughly justify the adoption of simply produced light-weight die-cast components in place of the frequently heavier and always more cumbersome and expensive composite assemblies which were formerly considered indispensable for such purposes.

## Institute of Metals

THE London Section of the Institute of Metals announce a meeting on September 4th, at the Junior Institution of Engineers, 39, Victoria Street, Westminster, S.W.1, at which Dr. W. D. Jones will lecture on "Developments in Powder Metallurgy."

# Intergranular Corrosion in Austenitic Stainless Steels—Part II.

By J. H. G. Monypenny, F.Inst.P.

*The occurrence of intergranular corrosion in austenitic chromium-nickel steels, its detection, cause and prevention and the effects it produces, are discussed. In the practical application of these steels this trouble has been largely overcome, though not entirely, and in this article the author takes stock of the present position, studies in some detail the effect of different variables, and points out where satisfactory results can be obtained in the industrial use of the steels and where further improvements are still desirable. Attention is given mainly to the 18-8 steels, as these are the most widely used of the austenitic stainless steels, but the author emphasises that intergranular corrosion is not confined to steels of this composition.*

**I**N the first part of this article detection of intergranular susceptibility by microscopic examination was described. For general use, however, a form of corrosion test which indicates definitely the presence or absence of intergranular attack, and is not dependent on the skilled microscopic examination of a carefully etched sample, is much to be preferred. The corrosive reagent most frequently used in this country (and often in other countries) for such a test is that originally developed at the Brown-Firth Research Laboratory under Dr. Hatfield, and well-known owing to its adoption as a standard method of test by the Air Ministry as well as by many large private concerns. The reagent consists of an aqueous solution containing 111 grams of copper sulphate crystals and 98 grams of concentrated sulphuric acid (density 1.84) per litre.<sup>‡</sup> It is used at boiling point, the period of immersion being generally 72 hours; this period has, in fact, been standardised in Air Ministry and other specifications. The reagent attacks rapidly an "18-8" steel which has been made susceptible by faulty treatment, but has no action on one that has been properly heat-treated. In the latter respect it is superior to a number of other reagents whose use has been proposed at various times, and which include boiling concentrated nitric acid (density 1.4), hydrochloric acid, hydrofluoric acid and various mixtures of these acids, all of which attack the properly treated steel to a greater or less extent.

When attack occurs with the acid copper sulphate reagent, copper is deposited on the steel sample, thus reducing the copper sulphate content of the solution. If this content is reduced considerably, the acid liquor will commence to attack the properly treated steel; hence, when tests are being made, there should always be an ample volume of the solution, and if a considerable amount of metallic copper is precipitated—owing to attack on one or more samples—it may be advisable to change the solution during the progress of the 72-hour test.

It is well known that the absence of attack on properly heat-treated material is due to passivity produced by the copper sulphate. The standard reagent actually contains considerably more of the latter salt than is required to produce passivity in an austenitic steel containing 16 or 18% chromium, but barely enough to have the same effect on one containing only about 12 or 14%. When such low chromium steels are being tested it may be advisable to increase somewhat the copper sulphate content of the solution. In the same way, a reduction of this content should tend to make the reagent still more sensitive in its action. The author showed some years ago<sup>13</sup> that the addition of copper sulphate to dilute solutions of sulphuric acid inhibited the attack of the latter on stainless irons, and that the amount of copper sulphate necessary to produce passivity against a 25% solution of the acid at atmospheric temperature was lower, the higher the

chromium content of the stainless iron. Probably the chromium content of austenitic stainless-steels has a similar effect with boiling 10% solutions of the acid. Rutherford and Aborn<sup>14</sup> used a solution containing 47 cc. concentrated sulphuric acid (density 1.84) and 13 grs. crystallised copper sulphate per litre, and claimed that its action was faster than that of the 10% solution commonly used; hence, they reduced the period of attack to 24 hours. They remark, however, that the copper content of this solution is insufficient to prevent general attack on "18-8" alloys, containing more than about 0.15% carbon, which have been subjected to prolonged carbide precipitation periods. Presumably such attack would result from the general lowering of the chromium content of the austenite brought about by fairly complete carbide precipitation and reasonable chromium diffusion from the grains to the depleted areas. As the precipitation of 0.15% carbon could thus result in the general lowering of the chromium content of the austenite by about 2%, it may be assumed that general attack would also occur in a properly treated steel containing less than about 16% chromium. The applicability of this reagent would thus appear to be limited strictly to alloys with a minimum of 18% chromium. In addition, a considerable volume of the reagent is necessary for tests on even small specimens so as to prevent any appreciable reduction in the copper sulphate content of the reagent during the progress of the test; actually Rutherford and Aborn recommend an amount of 250 cc. per square inch of surface of the sample, and they stress the importance of the sample being thoroughly clean and free from scale because the presence of oxide appears to induce excessive copper deposition. They also state that contact of flat surfaces of glass and steel appears to have a similar effect, hence flat specimens should be supported edgewise or on glass rods. Whilst it may be agreed that the Rutherford and Aborn reagent is more rapid in its action than the original Hatfield solution, it is also obviously much more sensitive to slight variations in testing conditions, and on that account is less suitable for use as a works inspection test to be applied to large numbers of samples.

Having subjected the sample to the selected corroding reagent, the next stage is to determine definitely whether or not intergranular attack has occurred. As is well known, the samples may be dropped on a hard wooden floor when the absence of the ringing sound characteristic of a good sample indicates marked intergranular attack. More definite evidence is obtained by bending the samples through 90° or so, when the depth of the cracks produced on the tension side of the bend indicates roughly the depth to which attack has penetrated into the sample. Inspection of such bent samples would allow them to be graded without much difficulty into four or five groups, ranging from a very slight attack affecting only a shallow surface

<sup>‡</sup> This reagent has been erroneously described on several occasions as the Strauss reagent.

<sup>13</sup> "Stainless Iron and Steel," 1st Edition (1926), p. 191.

<sup>14</sup> "A Quantitative Method for the Estimation of Intergranular Corrosion in Austenitic Stainless Steels," J. J. B. Rutherford and R. H. Aborn, *Trans. Am. Inst. Min. Met. Eng.*, Vol. 100 (1932), p. 293.



layer to complete penetration of the sample, and probably such a classification is sufficient for most purposes. It is certainly enough for practical testing in which the question to be answered is whether or not attack has taken place; if it has, the steel fails to meet requirements, and one is not greatly concerned with the rate at which failure occurred.

It may be useful, however, *e.g.*, in investigations regarding the effects of various treatments or of other alloys on susceptibility, to have a more accurate measurement of the depth to which attack has penetrated during immersion in the acid copper sulphate reagent. At first sight it might appear that this could be obtained by microscopic examination of a cross section of the corroded sample; actually, however, the surface of a badly attacked sample may be so loose that there may be difficulty in retaining it *in situ* during sawing, grinding and polishing, and its loss would also entail the loss of the datum line from which measurements of the depth of attack would be made.

A very accurate indication of the depth to which attack has penetrated may be obtained, however, by measuring the electrical resistance of the sample before and after corrosion, a method proposed by Rutherford and Aborn.<sup>14</sup> As intergranular attack leads to the removal of a thin layer of metal between two adjacent grains, the electrical resistance across this attacked boundary layer increases enormously; to all intents and purposes, current will be conducted through an attacked sample only by the unattacked core which remains. As the resistance of a sample varies inversely with the cross sectional area of this conducting core for a given length of the specimen, its measurement before and after corrosive attack provides a quantitative estimate of the extent to which intergranular attack has occurred. Probably wire is the most convenient form of sample for such a test, but sections of strip or sheet can readily be adapted, and from more massive samples test pieces about 3 in. long and of a uniform cross section (*e.g.*,  $\frac{1}{8}$ - $\frac{1}{4}$  in. diameter) may be machined. As the variable which is being measured is essentially the rate at which corrosive attack occurs, it is important, when a series of samples are being compared, that the opportunity for attack to occur should be similar for all the samples. Thus, they should all have substantially the same size and shape, and their surface finish should be the same. The latter may be achieved by some standard procedure in grinding, machining or etching the surface preparatory to immersion in the corroding medium. In their earlier tests, Rutherford and Aborn immersed their samples for a standard period of 24 hours in the special acid copper sulphate reagent which they developed and employed the percentage increase in resistance as an indication of the extent to which intergranular attack had taken place. Later, however, they calculated the actual depth to which attack had penetrated from the electrical resistance data, and as they found that this depth increased linearly with time in the corroding fluid, they converted this depth into penetration per 100 hours. The linear time relation enabled them to use short periods of attack (*e.g.*, two or three hours) for very susceptible samples (thus avoiding the difficulty of handling samples almost completely disintegrated), and much longer times for others only very slightly susceptible. The sensitiveness of the electrical resistance test enabled them also to detect and measure slight variations in the rate of attack practically indistinguishable under ordinary testing conditions.

The remarks made earlier regarding the suitability of the standard acid copper sulphate reagent for steels of chromium content lower than 18% inevitably raise the question as to the extent to which chromium depletion must progress before its presence can be detected by this reagent. If one assumes that the reagent is on the border line of attacking a properly treated steel containing about 13% chromium, it follows that depletion need not progress below this figure before the reagent will produce intergranular attack. Possibly it may not need to proceed as

far as this because the depleted area is sandwiched between carbide particles and grains of normal analysis, and to both of these it is likely to be anodic but, so far as the author is aware, evidence is not available as to the relative importance of such electrochemical action or to what extent the chromium content would have to be raised above 13%—the value assumed above—before such action became negligible; probably it will vary with the actual corroding medium. It seems very probable, however, that depletion must proceed below some value—perhaps about 15%—before appreciable intergranular attack is produced by the standard reagent, and it is conceivable that a somewhat smaller chromium impoverishment could be detected by the more sensitive reagent proposed by Rutherford and Aborn.

The train of thought thus started leads to a number of interesting questions regarding the practical use of these austenitic steels, and these will be discussed later in this article; it also has a bearing on the claims which have been made that intergranular attack can sometimes be developed by boiling concentrated nitric acid (density 1.4) in samples which are quite unattacked by the acid copper sulphate reagent. As is well known, boiling nitric acid of this strength attacks all commercial forms of stainless steel even if they contain as much as 28% chromium<sup>15</sup>, but the rate of attack increases quite appreciably as chromium content falls below about 18%. Thus Huey<sup>16</sup>, who suggested the use of this acid as a standard corrosion test for general purposes (*i.e.*, not in connection with intergranular corrosion), found the following rates of attack in stainless irons which were subjected to five 48-hour periods of immersion in the boiling acid:

Cr. %	Rate of attack (grams/m <sup>2</sup> /hour)
12.4	8.8
15.8	1.83
18.5	0.70

As these irons contained 0.1% carbon or rather less, it may reasonably be assumed that their effective chromium content (*i.e.*, the amount actually in solution in the ferrite) was about one per cent. less than the figures given; probably properly treated austenitic chromium-nickel steels having chromium contents equal to these amended figures would be attacked at similar rates. It is conceivable, therefore, that if a sample of "18-8" steel were heated for a very short time at 650° or 700° C., the incipient carbide precipitation so produced might reduce the chromium content at the grain boundaries to 16-17%, and that such a sample would not show any signs of attack when tested in the acid copper sulphate reagent, but would be attacked at a slightly faster rate by boiling concentrated nitric acid than a properly treated sample of the same steel. On the basis of the latter test, therefore, the slightly heated sample could be regarded as being not so resistant, as a properly treated sample, to boiling concentrated nitric acid or perhaps to some other acids—if such exist—whose rate of attack on stainless steels is related to chromium content in the same way, but it is questionable whether the test has any significance regarding the behaviour of the slightly heated sample under corrosive conditions which "18-8" steels resist with a reasonable margin of safety. The latter category would embrace the great majority of the purposes for which the austenitic stainless steels are used, and would even include the conditions in many synthetic nitric acid plants where resistance to the attack of this acid up to concentrations of 60% or so and at temperatures not above 85° C. is called for, and is met adequately by a minimum chromium content of 15% or 16%. For all such purposes, the acid copper sulphate reagent would seem to serve its purpose satisfactorily; in the few cases, mostly in the chemical industry, where liquors are concerned whose intensity of corrosive attack is somewhat similar to that of boiling

<sup>15</sup> See for example "Stainless Iron and Steel," 2nd Edition, Table LXXX., p. 351.

<sup>16</sup> "Corrosion Testing for Research and Inspection of Alloys," W. R. Huey, *Trans. Am. Soc. Steel Treat.*, XVIII, 1930, p. 126.

concentrated nitric acid, the use of the latter, or of the actual liquors involved, as a test reagent may give information as to susceptibility which the acid copper sulphate reagent would not disclose.

Before considering the second class of tests for intergranular corrosion, namely, those intended to determine whether a given steel is likely or not to become susceptible as a result of welding, or other fabrication process, or of conditions of use involving heating above atmospheric temperature, it may be useful to study the effect of different variables, e.g., time and temperature in the sensitizing range, prior treatment and composition, on the degree of susceptibility in different steels.

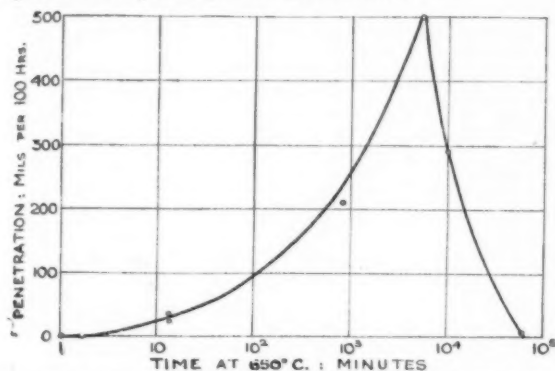


Fig. 7.—Rate of intergranular attack on "18-8" steel containing 0.08% C after heating for various periods at 650° C. (Bain, Aborn and Rutherford.<sup>17</sup>)

The picture, sketched in the first part of this article, of the course of carbide precipitation in an austenitic "18-8" steel, when held at 600° C., indicates that susceptibility develops more or less gradually. It becomes detectable after the steel has been held at some given temperature in the sensitizing range, for a period which may be very short—i.e., a matter of seconds—or considerably longer, becomes more and more pronounced with time, reaches a maximum and then diminishes, finally disappearing. The electrical resistance method of testing enables this sequence to be shown very strikingly; thus Fig. 7 gives the rates, as determined by Bain, Aborn and Rutherford<sup>17</sup>, at which intergranular attack penetrated into samples of an "18-8" steel, containing 0.08% carbon, which had been held for various periods at 650° C. In this case susceptibility developed moderately slowly, being hardly detectable after five minutes at this temperature, reached maximum

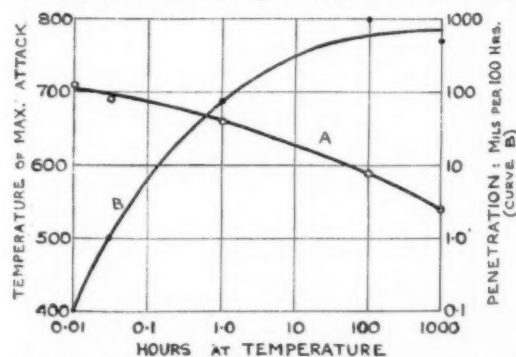


Fig. 8.—Periods of heating at various temperatures producing maximum susceptibility to intergranular corrosion and maximum rates of penetration so obtained. (Bain, Aborn and Rutherford.<sup>17</sup>)

intensity after about 90 hours and was removed, as a result of chromium diffusion, in 45 or 50 days. At higher temperatures—e.g., 700° to 750° C.—these investigators found that susceptibility developed rather more quickly,

but the maximum degree attained was lower than at 650° C., as a result, presumably, of more rapid diffusion of chromium. On the other hand, at lower temperatures, susceptibility was produced more slowly but reached a greater intensity before the very slow diffusion of chromium began to have an appreciable effect. This combined effect of time and temperature is shown graphically in Fig. 8, in which curve A gives the relation between temperature and time producing maximum susceptibility, and curve B the maximum rate of penetration so obtained. Curve A actually represents the mean value of a temperature range which extends 50° to 70° above and below the values plotted. It will be noted that with this particular steel, which contained 0.08% carbon, the curve representing maximum rate of attack appears to become horizontal at about 1,000 hours: Bain, Aborn and Rutherford mention in their paper<sup>17</sup> that tests carried out over periods up to 5,200 hours confirm that little, if any, greater susceptibility is produced if the heating period required to obtain maximum rate of attack is over 1,000 hours. The fact that the maximum degree of susceptibility falls as temperature rises—at least above about 550° C.—suggests that the rate of diffusion of chromium increases more quickly than that of carbon as the temperature rises. Even at the lower end of this temperature range, however, diffusion of chromium will occur in time and remove the intense susceptibility previously produced. Thus, periodical examination of low carbon "18-8" tubes built into an experimental superheater at the Trenton Channel plant of the Detroit Edison Co. showed that these tubes were very markedly susceptible after exposing for 267 and 605 days at 540° to 600° C., but became practically free from susceptibility after a total period of 900 days.<sup>18</sup>

In the practical use of these steels, however, one is generally not so much concerned with the maximum rate of intergranular attack which can be developed in a particular steel after heating for relatively prolonged periods as with the longest time it may be held in the sensitizing range of temperature without producing a measurable degree of susceptibility; this period of time is obviously a deciding factor in determining whether the steel is suitable for use in the welded condition without subsequent heat-treatment. This was the problem investigated by Rollason<sup>19</sup>, who treated a number of steels for various periods at different temperatures and then tested them, using the standard acid copper-sulphate reagent (72 hours' immersion) as the criterion of the presence or absence of intergranular susceptibility. His steels were generally of higher carbon content than that which formed the basis of Bain, Aborn and Rutherford's tests, the most complete data being obtained from one containing 0.19%. The results from strip  $\frac{1}{8}$  in. thick of this steel, which had previously been heated to 1,100° C., and then air cooled, are given in Fig. 9, and calculations from these data indicate that between 700° C. and 400° C. the time to produce a detectable amount of susceptibility is related to temperature according to the equation where

$$\theta = KT^m,$$

$\theta$  = temperature,  
 $T$  = time,

and  $m$  and  $K$  are constants having the following values:

$$m = -0.065,$$

$$K = 667,$$

the equation may be written:

$$\log \theta = m \log T + \log K,$$

indicating that when  $\theta$  and  $T$  are plotted logarithmically the results fall on a straight line, as shown in the graph. Data from steels of other carbon contents—with or without small additions of tungsten or copper—indicated that similar equations also applied to these steels, except that the value of  $K$  varied from steel to steel (i.e., the curves were displaced to a greater or less extent along the time

<sup>17</sup> "The Nature and Prevention of Intergranular Corrosion in Austenitic Stainless Steels," E. C. Bain, R. H. Aborn and J. J. B. Rutherford, *Trans. Amer. Soc. Steel Treat.*, Vol. XXI (1933), p. 481.

ordinate) and the equations held for a more limited range of temperature; below about 500°C., the time values observed were longer than corresponded to the equation, the curves tending to become asymptotic. It may be noted that if the equation held down to 450°C., the time required to produce a detectable amount of susceptibility at that temperature would be approximately 300 times as long as at 650°, whereas, as a result of the tendency mentioned above, the actual values recorded for some of the steels at 450° C. were 1,000/2,000 times as long as those for the same steels at 650° C.

Fig. 9 indicates that susceptibility was produced most quickly between 700° and 750° C., and that above the latter temperature the time required increased rapidly, a somewhat astonishing effect considering the high carbon content of the material.\*

Tests carried out in the Research Laboratory, Brown, Bayley's Steel Works, Ltd., on steels containing 0.11% to 0.16% carbon indicated that marked susceptibility was produced by heating them for 1, 2, 4, 8, 16, 32 and 64 mins. at 700°, 750°, 800° and 900° C., and probably these results are more representative of what is likely to occur in practice when such high-carbon steels are heated in this range of temperature. There are, however, a number of factors involved, and a consideration of these suggests that either type of result might be obtained, depending on conditions. In this temperature range, separation of carbide will occur as at lower temperatures, though perhaps at a somewhat slower rate as a result of the gradually increasing solubility of carbon in the steel. On the other hand, both carbon and chromium will presumably diffuse at a faster speed as the temperature rises, the increase in rate being probably greater with chromium than with carbon, as mentioned earlier. This action should tend to make the maximum development of susceptibility occur at an earlier stage as the temperature rises, as indeed Bain and his colleagues have found to be the case up to about 750° C. If one assumes that the limiting solubility at 850° C. is about 0.05%, as indicated by Bain, Aborn and Rutherford (curve A in Fig. 1, in the first part of this article†), and that the whole of the carbon in the steel tested by Rollason had been taken into solution during the previous softening treatment, and was evenly distributed through the austenite, the precipitation of the carbon atoms actually in the surface layers of the grains, and in excess of the solubility limit, would reduce the chromium content of these layers by about 2 per cent. As the steel contained only 15.2% chromium, this initial impoverishment should presumably make these layers susceptible to attack.

At these higher temperatures, however, there is an increasing tendency for the carbide to ball up into disconnected particles instead of remaining as a continuous film, and once this has occurred further carbide will tend to precipitate on the balled-up particles rather than in the grain boundary space between them. It seems probable

that the speed at which the sample is heated will have a marked effect on the rate at which coalescence will subsequently occur. A sample heated very rapidly—e.g., a thin strip immersed in a bath of molten tin or lead, previously heated to the testing temperature, the method used by Rollason—will arrive at this temperature without any appreciable precipitation having occurred at lower temperatures during heating, whereas one heated more slowly—e.g., in an ordinary muffle furnace—will have time to develop a distinct carbide network before it reaches the testing temperature. The writer used a molten salt bath, which gives a rate of heating probably faster than a muffle but slower than a molten metal-bath. A very thin film of carbide balls up much more readily than one considerably thicker, hence one may assume that a film produced initially at, for example, 850° C. in the rapidly heated sample will start to coalesce immediately it is formed, with the result that an appreciable time may elapse before a sufficiently marked continuous depleted layer is produced at the grain surfaces; in fact, such a layer, under favourable conditions, may not be produced at all. On the other hand, the film formed in a sample during slow heating will have produced a continuous layer, appreciably impoverished in chromium, at the grain boundaries before the steel reaches a temperature high enough to cause marked coalescence. Such a sample will already be susceptible before it reaches 850° C., and the early stages of its sojourn at that temperature will very probably increase susceptibility, because the film already formed during heating is too thick to ball up immediately. How long a period at 850° C. will elapse before removal of susceptibility begins will depend on circumstances, but one can imagine that a sturdy carbide film, developed by moderately slow heating, such as would occur with massive pieces of steel, would remain intact until carbide precipitation was well-nigh complete, in which case removal of susceptibility would await adequate chromium diffusion to enrich the heavily depleted layers.

As thick films would be the more readily produced during hearing, the higher the carbon content of the steel, one may expect that the difficulty of obtaining quickly a balled-up structure, together with relative freedom from susceptibility, will increase with carbon content, other things being equal. A further corollary is that the temperature range in which susceptibility is produced will extend higher, the greater the carbon content of the steel. As some evidence of this, it may be noted that steels containing 0.06% to 0.08% carbon when heated for 1, 2, 4, 8, 16, 32 and 64 mins. at various temperatures, in the same way as the 0.11% to 0.16% carbon steels mentioned above, were quite free from intergranular weakness at 900° C., and only developed slight susceptibility at 800° C. These points will be referred to later in connection with their practical aspect.

Preferential precipitation on existing carbide particles will also occur if, during previous softening, all the carbide in the steel had not been taken into solution. In addition, the carbide thus remaining undissolved reduces the amount of carbon available for intergranular precipitation, the steel thus behaving during subsequent reheating at, e.g., 650° C. as if it were of lower carbon content. It seems possible that this may have been an important factor in the relative freedom from intergranular corrosion, which Rollason obtained at temperatures over 800° C. and shown in Fig. 9.

Some of the earlier investigators on intergranular corrosion do not seem to have fully appreciated the effect of time at the sensitizing temperature in gradually producing more and more intense susceptibility up to a maximum value followed, with still longer periods, by a gradual recovery, susceptibility being finally entirely removed. Different investigators exposed their steels for widely

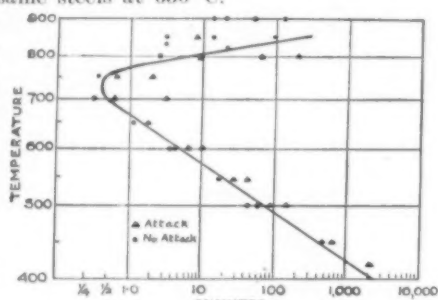


Fig. 9.—Time of heating at various temperatures required to produce susceptibility to intergranular attack in steel containing 0.19% carbon, 15.2% chromium, 10.9% nickel (Rollason.<sup>19</sup>)

\* It may be noted that one result at 850° C. did not agree with the rest of the data at that temperature. The curve as drawn by Rollason gives 300 mins. as the approximate minimum time to produce susceptibility at this temperature, the position of the curve being mainly determined by results at 800° and 825° C. Four samples were actually tested at 850° C., and of these, one heated for 8 mins. suffered from intergranular attack. The other three heated, respectively, for 3½, 15 and 90 mins., were not attacked.

† July, 1941, METALLURGIA.

<sup>18</sup> "High Temperature Steam Experience," P. W. Thompson and R. M. Van Duzer, *Combustion*, Nov., 1933.

<sup>19</sup> "Intergranular Corrosion of '18-8' Austenitic Stainless Steels," E. C. Rollason, *J. Iron and Steel Inst.*, 1933, 1, p. 391.



different periods at the sensitizing temperatures; thus, Payson<sup>20</sup> adopted 10 mins., Miller<sup>21</sup> 20 hours, and Becket and Franks<sup>22</sup> a month. It is not surprising that agreement was not always obtained as to the effect of composition or prior treatment on susceptibility. The period adopted by Payson (10 mins.) was intended to reproduce conditions occurring in ordinary welding operations, and although doubtless it is rather short when the welding of thick plate, involving several "runs," is in question, or the welded article is complex, necessitating several welds meeting at one place, the data he obtained are useful from this point of view. On the other hand, periods of 20 hours or a month have little or no relation to such practical questions, and their use is quite likely to brand as "susceptible to intergranular corrosion" steels which could quite safely be used in the welded condition without subsequent heat-treatment. Miller<sup>21</sup>, for example, came to the astonishing conclusion that "intercrystalline corrosion in an alloy is about as bad if the carbon content is 0.06% or 0.15%, the only advantages of the low carbon being the more-restricted temperature range in which precipitation occurs." One can agree that after heating for 20 hours at 600° to 650° C., both steels would be markedly susceptible to intergranular attack—though the higher carbon steel would almost certainly be worse than the other,—and that susceptibility would be produced, with the same period of heating, at higher temperatures in the higher carbon steel, but Miller failed entirely to appreciate that susceptibility developed very much more slowly in the low carbon steel, and it is to this fact that such steels largely owe their usefulness in practice.

Results obtained from these long-period heatings may be misleading in another respect. Heating for a month at the upper end of the sensitizing range of temperature would inevitably lead to complete recovery in many steels, and hence results so obtained, unless accompanied by data from shorter periods of heating, are apt to give an erroneous impression of the range of temperature producing susceptibility in some particular steel. This is the case, for example, with the results in the following table which were obtained by Becket and Franks, and are quoted with little comment and certainly no qualifying explanation in the second volume recently published of "The Alloys of Iron and Chromium."<sup>23</sup> In thus implying that disintegration is not producible in such steels, or at least those containing 0.07% carbon and upwards, at temperatures higher than are indicated, this table is very misleading.

SENSITIZING TEMPERATURE RANGE OF CHROMIUM-NICKEL STEELS water-quenched from 1,150° C. The steels were held at temperature for one month. (Becket and Franks.)

Composition.			Temperatures causing disintegration on boiling for 200 hours in acidified copper-sulphate solution.
Carbon, %	Chromium, %	Nickel, %	
0.03	12.74	6.27	400°—550°
0.04	12.82	12.80	"
0.05	12.52	25.44	"
0.04	18.23	8.49	"
0.07	17.47	9.27	400°—650°
0.12	18.45	8.56	400°—750°
0.19	18.40	9.37	"
0.09	18.37	14.12	400°—550°
0.09	20.42	11.87	"
0.04	22.25	15.06	475°—550°
0.05	22.32	20.38	"
0.07	27.32	23.19	"

The conception of a "safe period," representing the maximum time a given steel may be heated in the range of temperature which rapidly induces susceptibility—650° C. is convenient for testing purposes—without developing any detectable amount of intergranular weakness, is very useful in the practical application of the steels. It has a

reasonably constant value for a particular cast of steel, however, only if the prior treatment of the steel has been the same or, rather, has been such as to produce a similar final structure. Variations in grain size and complete or only partial solution of carbide are two main factors likely to affect susceptibility. Incomplete solution has already been referred to; it reduces the amount of carbon available for producing an intergranular network, and, in addition, provides nuclei on which innocuous precipitation may occur, instead of dangerously at grain boundaries. If incomplete solution is due to insufficient time at the heat-treatment temperature to allow equilibrium to be reached, this innocuous precipitation may be increased because, in such circumstances, it is unlikely that the carbon actually in solution will have diffused evenly through the austenite. Its concentration will probably be greater round the carbide particles which remain undissolved, with the result that preferential precipitation on the latter will be assisted during subsequent heating in the sensitizing range of temperature. Sheet and strip are rarely held for more than a few minutes at the softening temperature, in order to avoid undue scaling; if equilibrium has not been attained, such material will tend to give a longer "safe period" than if it had been more completely softened.

Low softening temperatures produce a fine grain size in worked material, and this also delays the development of susceptibility. A fine grain size means a greatly increased area for precipitation of carbide, hence a much smaller density of precipitation, and therefore a longer time to

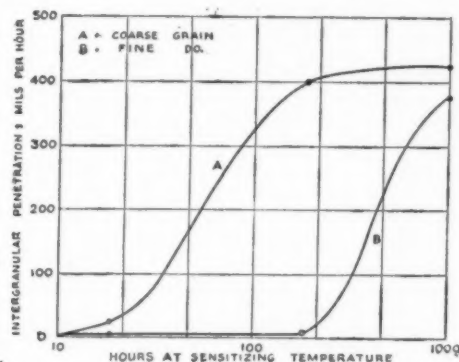


Fig. 10.—Relative development of susceptibility in coarse- and fine-grained material containing 0.05% carbon, 17.6% chromium, 8.7% nickel. (Bain, Aborn and Rutherford.<sup>17</sup>)

produce a degree of chromium impoverishment, which will cause detectable susceptibility. Fig. 10 gives results obtained by Bain, Aborn and Rutherford<sup>17</sup> on the development of susceptibility in fine and coarse-grained samples of "18-8" steel containing 0.05% carbon. The difference between the two curves is very marked; it would probably be less with steels of higher carbon content. As has been pointed out by Newell,<sup>24</sup> however, it is often sufficient to cause a marked difference in the behaviour of a given steel in service.

The effects of prior treatment make it essential, however, that in comparative tests as to the "safe periods" of different steels the latter shall have had comparable treatment, and if the results are to be applied in practice this treatment should consist of softening at a temperature, and for a time not less than will actually be used for the commercial softening of the steel. When plant requiring complex welding operations, including numbers of places where two or more welds meet, is in question, it is an advantage also to determine what effect softening at 1,250° C., for example, has on the "safe period" of the steel.

<sup>20</sup> "Prevention of Intergranular Corrosion in Corrosion-Resistant Chromium-Nickel Steels," P. Payson, *Trans. Amer. Inst. Min. Met., Eng.*, Vol. 100 (1932), p. 306.

<sup>21</sup> "An Investigation of certain Corrosion-Resistant Steels," J. L. Miller, *Carnegie School Mem., Iron and Steel Inst.*, 1932, p. 11.

<sup>22</sup> "Effects of Columbium in Chromium-Nickel Steels," F. M. Becket and B. Franks, *Trans. Amer. Inst. Min. Met., Eng.*, Vol. 113 (1934), p. 143.

<sup>23</sup> "The Alloys of Iron and Chromium, Vol. II," A. B. Kinzel and B. Franks, McGraw Hill Publishing Co., Ltd. (1940).

<sup>24</sup> "Influence of Grain Size on the Properties and Corrosion Resistance of the 18-8 Iron-Chromium-Metal Alloy for Elevated Temperature Service," *Trans. Amer. Soc. Steel Tre-t., Vol. XIX*, 1931-1932, p. 673.

# The Influence of Crystal Structure on the Cold Rolling and Annealing of Copper Strip

By Maurice Cook, D.Sc., Ph.D., and T. Ll. Richards, B.Sc., Ph.D.

*A study has been made by X-ray and microscopic methods of the effect of progressively increasing cold-rolling reductions and of subsequent annealing on h.c. copper strips having four different types of crystal structure. The results of this study are recorded in a paper\* presented to the Institute of Metals for discussion, and are reproduced here in a substantially abridged form.*

THE effect of progressively increasing cold-rolling reductions and of subsequent annealing upon the structure of copper strip composed of randomly-orientated crystals of a particular grain size has already been studied by the authors.<sup>1</sup> It was shown that in strip cold-rolled with reductions in thickness of less than about 50%, no preferred orientation of crystals was present after annealing. It was also shown that with reductions in excess of 50% a twin-fibre texture was developed in the cold-rolled strip; this either recrystallised independently on annealing to form a double texture with two sets of crystals having preferred orientations corresponding to one or other of the twin-fibre textures, or the twin textures coalesced on annealing to form a single texture, the crystals having a preferred orientation such that a {100} plane was parallel to the strip surface and a <100> axis was in the rolling direction.

It is possible for rolled and annealed copper strip to have:

- a randomly orientated structure (type N);
- a structure composed of a mixture of randomly-orientated and double-texture structures;
- a double-texture structure (type D);
- a structure composed of a mixture of double- and single-texture structures (type M);
- a single-texture structure (type S).

Which structure or structures are present and the proportions in which they occur are determined entirely by the rolling and annealing conditions employed in the preparation of the strip.

The purpose of the work described in the present paper was to study the effect of progressively increasing cold-rolling reductions and subsequent annealing on copper strip composed of three different crystal structures, and commercial copper of h.c. quality, composed of 99.95 copper, 0.04 oxygen, 0.003 silver, and less than 0.005 iron, was prepared with structures of types N, D, M, and S. In the strip with the type M structure, double- and single-texture crystals were present in about equal proportions. Strip with each of these four types of structure was prepared in three grain-sizes by suitably adjusting the temperature of the last annealing operation, making a total of 12 different conditions of structure and grain-size in the basic copper used in the investigation.

The type of structure present in the annealing strip before the rolling and annealing treatments dealt with in this paper, is represented, in the marking of the samples for reference, by the initial letters N, D, M, or S. The grain-size in the strips at this stage is indicated by a second letter, A, B, or C. This letter is followed by a figure which indicates the cold-rolling reduction to which the strip was afterwards subjected, and the final figure indicates the temperature of the annealing which followed. Thus, S B 64 indicates a strip which had a single texture with a grain-size of 0.035 mm., and which was then cold rolled with a reduction in thickness of 90%, and subsequently annealed at 700°C. Similarly, a sample with only one figure following the two letters indicative of the original condition of the strip is one in the cold-rolled condition. The key to the numbering of samples is given

in Tables I and II, the letters giving the grain-sizes of the strips before rolling, which were all comparatively uniform, except the D B and C D sets, and for these the figures given are average values which cover a considerable variation in size.

TABLE I.

KEY TO NUMBERING AND LETTERING OF SPECIMENS.		1 Rolling reduction of 2.5%	
N	Randomly orientated structure.	2	" " 10%
D	Double-texture structure.	3	" " 25%
M	Mixture of double- and single-texture structure.	4	" " 50%
S	Single-texture structure.	5	" " 75%
1	Annealing temperature 400°C.	6	" " 80%
2	" " 500°C.	7	" " 85%
3	" " 600°C.	8	" " 90%
4	" " 700°C.		" " 92.5%
5	" " 800°C.		" " 95%
6	" " 900°C.		" " 97.2%
7	" " 1,000°C.		

TABLE II.

GRAIN-SIZES (IN MM.) BEFORE ROLLING AND ANNEALING.

	1.	2.	3.	4 & 4r.	5r & 5.	6r, 6, 7r, & 7.	8.
Structure N:							
A	0.03	0.025	0.02	0.025	0.025	0.03	0.03
B	0.04	0.04	0.04	0.035	0.04	0.04	0.04
C	0.06	0.06	0.06	0.065	0.065	0.065	0.065
Structure D:							
A	0.015	0.02	0.015	0.015	0.02	0.02	0.02
B	0.035	0.035	0.035	0.035	0.035	0.035	0.035
C	0.10	0.10	0.10	0.09	0.11	0.10	0.10
Structure M:							
A	0.03	0.025	0.02	0.025	0.025	0.025	0.025
B	0.04	0.035	0.035	0.04	0.035	0.035	0.035
C	0.06	0.065	0.06	0.06	0.065	0.065	0.065
Structure S:					5r, 5, 6r, 6, 7r & 7		
A	0.015	0.02	0.02	0.02	0.02	0.02	0.02
B	0.035	0.035	0.035	0.035	0.035	0.035	0.035
C	0.06	0.06	0.065	0.06	0.065	0.065	0.065

## Effect of Cold Rolling

The effect of cold-rolling is discussed following X-ray and microscopic examination and certain mechanical tests carried out on the various series of strips. The X-ray examination revealed that, in general, strips of the four series behaved in a similar manner during cold rolling and the deformation during rolling reductions up to about 50% took place by a slip mechanism. The same twin-fibre texture was developed by the heaviest cold-rolling reductions, although the single-texture orientation in the S and M series showed a tendency to persist. The degree of preferred orientation in the strip of the four series was independent on the annealing temperature or grain-size before rolling, and was dependent only on the type and degree of preferred orientation at that stage and on the magnitude of the subsequent rolling reduction. Partial and even complete recrystallisation of the twin-fibre texture was found to have occurred in those samples of the N, D, and M series with the smallest of these ranges of grain-size considered, and which had been subjected to rolling reduction of about 90% and more.

The effect on the microstructure of progressively-increasing amounts of cold-rolling appears to be the same in material of all four series. Up to reductions of about 50% the chief visible effects on the microstructure are an elongation of the crystals and a tendency for twin crystals to become orientated in the rolling direction. Microstructures do not provide an indication of the mechanism

\* Jour. Inst. Metals, 1941, 37, 203.

1. M. Cook and T. Ll. Richards, J. Inst. Metals, 1940, 66, 1.

of the deformation when the reductions are less than about 50%, but the X-ray evidence indicates that it occurs by a slip mechanism. After reductions of about 50% striations transverse to the rolling direction, which, as previously<sup>1</sup> noted, are evidence of a second mechanism of deformation, appear across the crystals. In all four series, striations of this type occur in about a quarter of the number of crystals after 50% reduction, in about half the number of crystals after 60% reduction, and in practically all of the crystals when the rolling reduction exceeded 80%. This effect on the crystal structure during rolling reductions in excess of about 50% was uninfluenced by preferred orientation of the crystals or their grain-size before rolling. In spite of the break-up of crystals by block movement, they still retain their individuality after rolling reductions of the order of 90%, even though they have at that stage become very elongated and extremely thin.

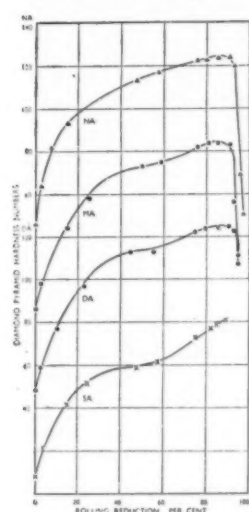


Fig. 25.

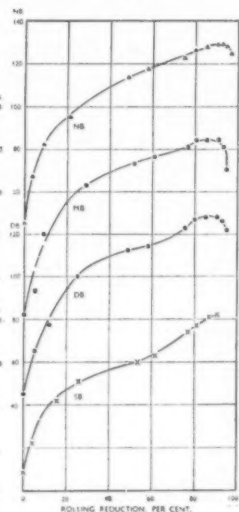


Fig. 26.

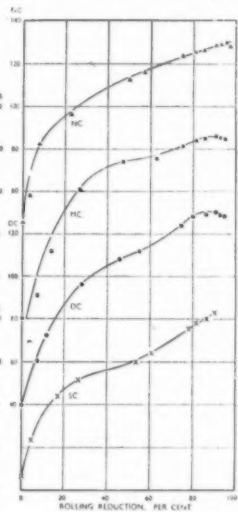


Fig. 27.

Diamond pyramid hardness tests of samples after various rolling reductions.

Diamond pyramid hardness tests were taken on all samples after the various rolling reductions, and the results obtained are illustrated in curves reproduced in Figs. 25, 26, and 27. Each figure shows four work-hardening curves for strip with the four initial structures *N*, *M*, *D*, and *S*, and each separate figure relates to the three ranges of initial grain-size considered. The chief features of these curves are that there is a progressive change in shape according to the initial structure and in some instances there is evidence of pronounced softening after higher rolling reductions. The softening effect is most marked in the curves in Fig. 25, which relate to strip with the smallest grain-size; it is less marked in Fig. 26 for material with an intermediate grain-size, and is least marked in Fig. 27 for material with the largest grain-size. With reductions up to the order of 90%, there is no evidence of any softening in the *S* material, the initial structure of which was of the single-texture type. These curves are based on hardness determinations carried out some weeks after the actual rolling; further tests which were made later showed that the softening did not occur during rolling, but took place subsequently at room temperature. Thus, material of the *NA* type, for example, showed no softening when tested immediately after rolling and strip with a cold-rolling reduction of 97% had a D.P. hardness of 127. This rolled strip progressively softened at room temperature, and over a period of months the hardness fell to 54. These observations provide an explanation of the anomalous results which have been recorded at different times by other investigators<sup>2</sup> concerning the effects of heavy rolling reductions on the hardness of copper. This self-annealing effect, which it will be seen is related to the magnitude of the cold-rolling

reduction, the initial grain-size of the material, and its original structure, is being further investigated.

The evidence from these hardness tests indicates that the recrystallisation effect, which was noted by X-ray methods in some of the samples and which has been commented on in a previous section of the paper, is due to the material annealing at room temperature after rolling. Evidence in some of the samples of this spontaneous annealing at room temperature was also provided by microscopic examination.

Tensile tests were carried out on strips after progressively increasing amounts of cold rolling, on standard (B.S.S. No. 465—1934) test-pieces cut at 0°, 45°, and 90° to the rolling direction. It was not found that variations in initial grain-size over the range considered had any marked effect on the tensile or elongation values, and figures are, therefore, reproduced only for the *B* set of strips, that is, those of intermediate grain-size, 0.035 to 0.04 mm.

From the results given in Table III it will be noted that *N* strip, which has a random orientation of crystals, shows, in the soft condition, no evidence of directionality. With increasing rolling reductions the tensile strength assumes a definite maximum in the transverse direction and a minimum value at 45°, a finding which confirms that reported by Phillips and Bunn.<sup>3</sup> With both 70:30 and 64:36 brass also, one of the present authors<sup>4</sup> has shown that when the rolling reduction has been of a magnitude sufficient to induce a directional effect, the maximum tensile strength is at a maximum in the transverse direction.

In the *DB* strip in the soft condition the tensile strength is at a maximum in the rolling direction, the values in the other two directions being less than normal. As with the *NB* strip, the effect of progressively-increasing rolling reductions is to develop a maximum value for tensile strength in the transverse direction, and a minimum value at 45°.

TABLE III.  
TENSILE TESTS ON COLD-ROLLED STRIP.

Reduction in Thickness by Cold Rolling, %	Direction of Test-piece Relative to Rolling Direction, Degrees	<i>N B.</i>		<i>D B.</i>		<i>S B.</i>		<i>M B.</i>	
		Tensile Strength, tons/in. <sup>2</sup>	Elongation, % on 2 in.	Tensile Strength, tons/in. <sup>2</sup>	Elongation, % on 2 in.	Tensile Strength, tons/in. <sup>2</sup>	Elongation, % on 2 in.	Tensile Strength, tons/in. <sup>2</sup>	Elongation, % on 2 in.
		Rolling Direction, Degrees		Rolling Direction, Degrees		Rolling Direction, Degrees		Rolling Direction, Degrees	
0	0	14.6	48	14.8	45	11.7	24	14.6	47
	45	14.1	49	13.8	49	12.6	74	14.1	53
	90	14.6	48	13.9	47	11.5	26	13.8	47
10	0	16.3	31	16.9	26	12.4	6	16.0	28
	45	15.9	36	15.8	31	13.1	58	15.1	41
	90	16.0	32	16.0	35	12.6	8	15.2	28
25	0	18.3	6	20.0	5	13.8	2	18.1	7
	45	18.1	5	19.4	5	11.1	32	17.6	6
	90	18.6	7	20.3	5	14.0	2	18.0	8
50	0	24.3	2	23.7	2	15.6	2	23.4	2
	45	24.0	2	23.2	2	15.6	3	22.9	2
	90	25.7	2	25.1	2	16.3	2	24.5	2
80	0	27.6	2	27.9	2	19.7	2	26.3	2
	45	27.3	2	26.9	2	17.2	2	26.1	1.5
	90	28.8	2	28.1	2	20.4	2	27.4	2
90	0	28.5	2	28.3	2	20.9	2	28.1	2
	45	27.9	1.5	27.6	2	18.2	2	27.2	2
	90	29.5	1.5	29.2	1.5	22.0	2	29.2	2
97.2	0	30.1	3	—	—	—	—	—	—
	45	27.9	2	—	—	—	—	—	—
	90	30.2	2	—	—	—	—	—	—

2. H. Moore, *J. Inst. Metals*, 1924, **32**, 407.

3. A. Phillips and E. S. Bunn, *Trans. Amer. Inst. Min. Met. Eng.*, 1931, **93**, 353.

4. M. Cook, *J. Inst. Metals*, 1937, **60**, 159.



The *SB* strip showed the most pronounced directional differences of all four types considered. It will be seen from Table III that in the soft condition the tensile strength in all three directions was much below the normal values for copper strip, and that the highest value was in the 45° direction. The elongation value was also at a maximum in this direction and, moreover, the value was much higher than normal, while in the other two directions, it was much below normal. As with the *NB* and *DB* strip, the effect of increasing rolling reductions was to develop a maximum tensile strength in the transverse direction and a minimum value in the 45° direction. It will also be noted that the tensile-strength values after any given reduction are much less for the *SB* strip than for strip of the other three types.

The *MB* strip, the structure of which is a combination of the *S* and *D* structures, showed, like the *DB* strip, a maximum tensile strength in the rolling direction and, like the *SB* strip, a maximum elongation in the 45° direction, features which are shown by strip in the soft condition and after a 10% rolling reduction. With increasing rolling reductions the same directional effect was developed as in the other series.

#### Effect of Annealing after Cold Rolling

**X-Ray Structure.**—Systematic X-ray examination was confined to strips of the four series which had been annealed at 400° C.; these had a relatively small grain-size and yielded diffraction patterns which could be interpreted more positively than those of strips at higher temperatures and with, in consequence, larger grain-sizes. While the observations recorded in this section, therefore, relate only to material annealed at 400° C., the results of microscopic examinations of strips annealed at higher temperatures indicate that the conclusions apply also to them.

The examination showed that the three sets of strips of the *N* series, with rolling reductions of up to 50%, recrystallized on annealing with a random orientation of crystals, except specimens *NB* 11 and *NC* 11 which had not completely recrystallized. With increasing rolling reductions above 50%, there was, after annealing, an increasing degree of preferred orientation of the crystals in all three sets of strip of this series. Two types of texture were developed: first, the double texture where the twin-fibre textures of the cold-rolled strips recrystallized independently, and second, the single texture where the twin textures had coalesced on annealing. The transition from one to the other was gradual; it occurred after greater rolling reductions as the annealing temperature before rolling was increased.

In the *D* series, specimen *DC* 11, like *NB* 11 and *NC* 11, had not fully recrystallized; apart from this one strip there was no evidence that the preferred orientation of the double-texture type, which characterized these strips before rolling, remained after rolling and annealing. The coarsening of the crystal structure brought about by the relatively small amount of deformation and subsequent annealing may, however, mask the presence of any preferred orientation. As in the *N* series, strips of the *D* series, with lower rolling reductions, showed, after annealing, a random orientation of the crystals. The presence, before rolling, of the double-texture structure does, however, influence the structure developed in the finally annealed strip. Thus, in *DA* 31, there is present preferred orientation of both *D* and *S* types; with increasing rolling reductions, however, the preferred orientation developed on annealing becomes more marked, until in annealed strip with reductions of the order of 80% or more it is wholly of the single-texture type. The *B* set of the same series shows similar features, except that the simultaneous presence of the two types of texture persists after greater rolling reductions and only after a reduction of about 95% is the structure entirely of the single-texture type. The *DC* strips had a coarse and irregular crystal structure before rolling, and photographs of them show that one or other of the twin textures predominates as in the corres-

ponding cold-rolled strips. There is also no indication in the *DC* set of annealed strips that any of the double-texture structure coalesced on annealing to form a single-texture structure even after a rolling reduction of 95%.

The diffraction patterns of *SA* samples, in the final annealed condition, reveal that after rolling reductions of 2½% and 10% the strip had recrystallized on annealing with an orientation of crystals approximating to that of the original single structure. The grain-size of sample *SA* 11 was about 2 mm. and the X-ray diffraction pattern that of a single crystal. With a final reduction of 25% the orientation after annealing is quite random. All the structures of strip annealed after reductions of 50% to 80% are mainly of the double-texture type. With reductions of 80% or more there is increasing evidence of the presence of single-texture structure. Variations in structure of the *SB* and *SC* strips with increasing rolling reductions are, in general, similar to those in the *SA* series, except that the structure of *SB* 11, *SC* 11, and *SC* 21 is that of a distorted single texture,—i.e., these strips have not completely recrystallized, and even in strips with rolling reductions of 90% to 95% only closely aligned double-texture crystals are present. The predominance of the double-texture structure in this series is a notable feature in view of the fact that the single texture appears to be so stable on rolling.

The *M* series, as might be expected, showed features of the other three. Strips *MB* 11, *MC* 11, and *MC* 21 were not completely recrystallized, and strips with heavier reductions recrystallized to a random orientation. In those with rolling reductions of the order of 50% or more, the double texture was present after annealing, and the preferred orientation became more pronounced at the highest reductions with a development of the single-texture type of structure at the expense of the double. As with the strip of the *N* series, and *DA* and *DB* strip, the transition from the double to single texture occurred after greater rolling reductions as the annealing temperature before rolling was increased.

**Microstructure.**—The strips of all the series, after the various rolling reductions followed by annealing at temperature intervals of 100° C. from 400° to 1,000° C., were examined microscopically, the structures being developed by an electrolytic method of etching with a solution about five times as strong as that recommended.<sup>5</sup>

The characteristic features of the different types of structure which can occur in rolled and annealed copper have been described in the section dealing with the structures of the strip before rolling and annealing; and all the strips of the series after the rolling and annealing treatments recorded showed structures covering these types. These can be recognised both by their etching characteristics and the twinning orientations. The latter provide an indication of both the nature and degree of preferred orientation, and one of the authors<sup>4</sup> has assessed the degree of preferred crystalline orientation in rolled and annealed brass strip by measuring the frequency of twinning in various directions.

The changes in structure in strip of all four series on annealing at 400° C., after different rolling reductions, as shown by X-ray evidence described in the previous section, were confirmed by micro-examination, the characteristic appearance of the single- and double-texture crystals making this possible. Similar changes in structure were noted in strips of all four series annealed at higher temperatures up to 900° C. Increasing the annealing temperature did not apparently influence either the degree or type of preferred orientation developed, but affected only the crystal size, which increased as the annealing temperature was raised.

Observations were made on the grain-size of strips of all four series with different rolling reductions followed by annealing at temperatures from 400° to 1,000° C., at 100° intervals. The smallest reduction employed—i.e., about 2½%, resulted in the usual coarsening of micro-

<sup>5</sup> C. H. Davis, *Metals Handbook* (Amer. Soc. Metals), 1939, 1471.

structure on subsequent annealing in strips of all four series, and particularly in the *S* series, except in the *B* and *C* strips, which did not recrystallise on annealing at 400° C. In strips of the *N*, *D*, and *M* series with this reduction the grain-size varied from about 0.1 mm. on annealing at 400° C. to between 1 mm. and 2 mm. on annealing at 1,000° C., whilst the grain-size of strips of the *S* series varied from about 2 mm. to about 2 cm. over the same range of temperature. This development of large crystals in material originally possessing a structure of the single-texture type is readily understandable, since the strip showed a pronounced preferred orientation of the crystals before rolling. These large crystals were characterized by the presence of a large number of twins at 45° to the rolling direction, which suggested that the nuclei for grain growth are distorted slip planes, and that the maximum grain growth takes place in the  $\langle 110 \rangle$  directions, corresponding to the closest packing of atoms in the face-centred cubic lattice.

Apart from the exceptions already noted, the grain-sizes in strips of all four series decreased with increasing rolling reductions up to about 50%, and, with any given

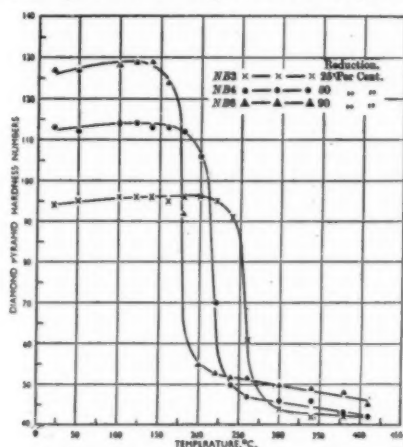


Fig. 28.—Showing the effect of variations in the magnitude of the final rolling reduction on the softening temperature of strip annealed at the same temperature before rolling.

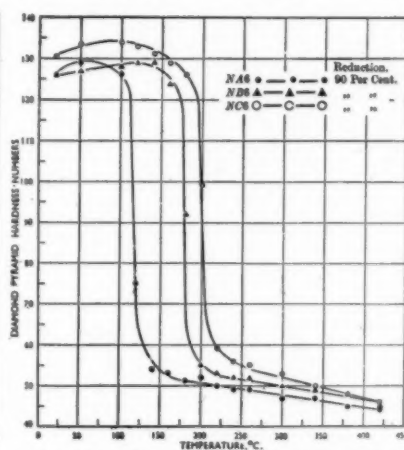


Fig. 29.—Showing the effect on the softening temperature of varying the annealing temperature, and in consequence the grain-size before rolling.

reduction, increase in annealing temperature resulted in the usual increase in grain-size without the occurrence of any abnormal grain growth. The random orientation of crystals developed by these conditions might be expected from the fact that with reductions of this magnitude the deformation which takes place by slip mechanism developed no marked preferred orientation.

The grain-size of strips with rolling reductions in excess of 50% and subsequently annealed at temperatures below 700° C. were generally relatively small and uniform, but with annealing temperatures in excess of 700° C., instances of abnormal grain growth occurred.

In some specimens only a comparatively small number of large equi-axed crystals occurred. These crystals, which ranged in size from about 1 mm. to 5 mm., generally appeared as isolated crystals dispersed in an otherwise uniform field of relatively small equi-axed crystals. The materials which showed this tendency to develop large equi-axed crystals on annealing at 700° C. or higher, were:

1. All strips of the *S* series with rolling reductions of about 50% or more, excepting *SA 6* and *SA 7*.
2. Strips of the other three series *N*, *D*, and *M*, annealed at 400° C. before rolling, and rolled with reductions of between 50% and 80%.
3. Strips of the same three series, annealed at 650° C. before rolling and rolled with reductions of between 50% and 90%.
4. Strips of the same three series, annealed at 900° C. before rolling, and rolled with reductions between 50% and 95%.

The above conditions yield strip which, in the annealed state, consists mainly of crystals of the double-texture type, and without exception the orientations of all the large equi-axed crystals observed correspond to one or other of the double-texture orientations. Observations indicate that whilst there is no tendency for abnormal grain growth to occur below 700° C. in such structures, much larger isolated crystals are formed at higher annealing temperatures.

Three specimens, *SA 57*, *MB 66*, and *MB 67*, were composed entirely of much larger equi-axed crystals. These ranged in size from about 2 cm. to 5 cm., and had irregular grain boundaries. X-ray and microscopic examination of the corresponding strips annealed at 400° and 600° C.—i.e., *SA 51*, *MB 61*, and *SA 53*, *MB 63*, showed that the structures were mainly of the double-texture type with a proportion of about 10% to 20% of the single-texture crystals. The large equi-axed crystals which characterise *SA 57*, *MB 66*, and *MB 67* were all of the double-texture type.

Of all the specimens examined six, *NB 87*, *NC 87*, *DB 77*, *MB 7x6*, *MB 76* and *MB 77*, were composed of large acicular crystals. All of the samples had a rolling reduction of at least 92½% and were annealed at temperatures of 1,000° C., except *MB 7x6* and *MB 76*, which were annealed at 900° C. No material which had originally the single-texture type of structure, nor any strips of the other three series annealed at 400° C. before rolling, developed these large acicular crystals, and they occurred only in one strip (*NC 87*) which had been annealed at 900° C. before rolling. Strips corresponding to those which developed large acicular crystals on annealing at 900° or 1,000° C., consisted, after annealing at lower temperatures, mainly of crystals of the single-texture type with a proportion of up to about 10% of double-texture crystals. Large acicular crystals were not developed in strips with an exclusively single-texture structure by annealing at 900° or 1,000° C. It appears, therefore, that to produce large crystals of the acicular type not only must the rolling reduction and final annealing temperature be very high, but also the initial conditions with respect to grain-size and structure must be such that the high rolling reduction followed by annealing at moderate temperature results in a structure with suitable proportions of crystals of both single and double-texture types.

**Physical Properties.**—The effect of variations in the magnitude of the final rolling reduction on the softening temperature of strip annealed at the same temperature before rolling is shown by the curves in Fig. 28, whilst the effect on the softening temperature of varying the annealing temperature, and in consequence the grain-size before rolling, is shown by Fig. 29. The curves were determined by taking the hardness of strips which had been heated for half-hour periods at temperatures ranging from room temperature to 420° C., the heating being carried out in an oil bath at temperatures up to 220° C., and at higher temperatures in a salt bath. For the determination of the curves in Fig. 28, samples *NB 3*, *NB 4*, and *NB 6* were used, that is, strips annealed at 650° C. before rolling, and cold rolled 25%, 50% and 90%, respectively. The strips used for obtaining the data for curves in Fig. 29 were *NA 6*, *NB 6*, and *NC 6*, that is, strips of the *N* series which had been given a rolling reduction of 90% after annealing at 400°, 650°, and 900° C., respectively. With a constant high rolling reduction of 90%, the soften-

ing temperature decreases with decreasing annealing temperature before rolling.

The softening temperature of *NA 6*, which had been annealed at 400° C., before rolling, was only about 120° C., and this material recrystallised by the coalescence of the cold-worked twin-fibre textures to form a single-texture structure. *NB 6* softened at about 180° C., and recrystallised to form a structure mainly of the double-texture type, but containing also some single-texture structure whilst *NC 6* softened at about 200° C. and the cold-worked twin textures recrystallised independently to give a double-texture structure. The results obtained with *NA 6* show that recrystallisation by coalescence of the twin-fibre textures to the single texture occurs at relatively low temperatures. This observation can be related to that previously noted concerning the self-annealing effects in those specimens of the *N*, *D*, and *M* series which were annealed at 400° C. before being subjected to very heavy cold-rolling reductions. Thus, it would seem that self-annealing can occur in material which possesses a relatively small grain-size before rolling, and which is rolled to such an extent that the twin-fibre textures recrystallise by coalescing to form the single texture. In connection with the influence of initial grain-size, if it is assumed that each grain of the original structure is rolled into a flat strip, the average thickness of the grains after a rolling reduction *R* can be calculated from the original grain-size. In Table IV are given the estimated values of the reduction *R* for the three sets of strips of the *N*, *D*, and *M* series and the calculated value *d* of the average thickness of the corresponding grain strips. From Table IV it can be seen in all cases that the values of *d* approximate to 0.003 mm.

Thus, when crystals are reduced by cold rolling to thicknesses of about 0.003 mm., or less, they can no longer recrystallise independently; this thickness is, therefore, a critical dimension for copper at which the twin textures of cold-worked strip coalesce on annealing to form a single texture.

TABLE IV.

Series.	Annealing Temperature Before Rolling.					
	400° C. (A).		650° C. (B).		900° C. (C).	
	<i>R</i> , %	<i>d</i> , mm.	<i>R</i> , %	<i>d</i> , mm.	<i>R</i> , %	<i>d</i> , mm.
<i>N</i>	~90	~0.003	> 92½ < 95	< 0.003 > 0.002	> 95 < 97.2	< 0.0035 > 0.002
<i>D</i>	~85	~0.003	> 92½ < 95	< 0.003 > 0.002	> 95	< 0.005
<i>M</i>	~85	~0.0035	> 92½ < 95	< 0.003 > 0.002	> 95	> 0.0035

Apart from specimens showing the abnormal grain-size already described, the variation in tensile properties of annealed material of all four series and the extent to which they show directional effects are illustrated by the first set of values in Table III, and reference has already been made in an earlier section to their chief features in relation to crystal structure. The presence of directional properties also shows itself by the occurrence of ears at the mouths of cups made from strip of all four series.

X-ray photograph illustrations of the specimens in each of the series are given in the paper for the material in the cold-rolled and the annealed conditions, while selected illustrations of the microstructure in different rolling direction also support the text recording this investigation

## A Deflectometer for Transverse Testing of Cast Iron

IN applying the transverse test to cast iron either for the control of foundry operation or for specification purposes, it is sometimes desirable to plot load deflection curves in order that resilience and relative modulus of elasticity may be determined. To do so takes some considerable time and is the work of two operators, so that if several hundred transverse bars are tested, the plotting of curves is a task of large proportions. An inexpensive method of obtaining autographic transverse load-deflection curves is highly desirable and such a method carried out by means of a deflectometer has recently been described by G. T. Greenidge and E. C. Kron.\* This new device gives a permanent record and requires only one person to make the test.

The deflectometer is based on the wedge principle, where the movement of a wedge, which is some multiple of the movement of the two halves of an extensometer, offers a simple means of measuring and multiplying the movement of the extensometer in the same operation. It consists essentially of a tube through which a rod slides and having one shoe fastened to the sliding rod and a second to the tube itself. Each shoe has two grooves in which the two wedges slide. When testing, the transverse bar rests on two supports attached to the moving head of the testing machine, and the rod of the deflectometer is held against the underside of the bar by a spring. During bending this rod remains stationary and the tube moves upwards, with the result that the shoe attached to the central and the one attached to the tube move apart. The two wedges which slide in the shoes move in a horizontal direction as the shoes move apart, because of the horizontal pull exerted on them by fish lines on which weights hang. One of the weights is fastened to one wedge by a line and a rubber

band, and because of the tension in the rubber band this wedge follows the rapid movement of the shoes when the transverse bar breaks and this prevents undue nicking of the other wedge which is attached to a recording drum. Nicking is suspended and, consequently, when the wedge moves the drum rotates, and since the wedge has a rise of 1 in. in 10 in., the linear movement of the recording drum is 10 times the deflection of the bar. The combined movement of a low-recording arm geared to the dial of the machine and of the drum cause a pen or pencil attached to the load-recording arm to trace a load-deflection curve.

The accuracy of the measurements obtained with this deflectometer are shown in Table I, where values obtained by a dial reading are compared with those obtained from autographic curves.

TABLE I.

Breakload, Lb.	Breaking Deflection in		Percentage of Error in Deflection.	Resilience, Ft.-lb.		Percentage of Error in Resilience.
	Dial.	Autograph.		Dial.	Autograph.	
0.875 in. BARS.						
1,550	0.143	0.143	0.0	10.6	10.6	0.0
1,540	0.134	0.133	-0.8	9.5	9.4	-1.0
1,880	0.152	0.151	-0.7	13.1	12.9	-1.5
1.2 in. BARS.						
2,600	0.235	0.236	+0.4	29.3	28.8	-1.7
2,820	0.234	0.232	-0.9	30.4	30.2	-0.6
2,790	0.233	0.249	+1.5	33.9	33.4	-1.5
2.0 in. BARS.						
8,700	0.317	0.320	+1.0	141.7	141.1	-0.4
8,470	0.299	0.303	+1.3	128.1	129.2	+0.9
8,650	0.287	0.293	+2.1	123.4	125.1	+1.4

In using the instrument, the wedges become slightly roughened from the impact of the shoes when specimens are broken. This slight roughening, however, had little effect on the accuracy of the values obtained, as was shown by taking a calibration curve on wedges which had been used in breaking 400 transverse bars of various sizes.

\* *Metals and Alloys*, 1941, vol. 13, 6, pp. 722-725.



# Developments in Continuous Annealing of Steel Strip

*The advantages possible by the adoption of continuous annealing applied to steel strip, as compared with present annealing practice, have caused much attention to be given to this operation in recent years. Reference is made to a few continuous furnaces which have been installed, but attention is directed more particularly to a vertical type of continuous furnace, equipped with a cooler, which has been developed, from which tinplate and similar thin material can be produced having characteristics making it suitable for many uses.*

**I**N recent years the manufacture of steel sheets and strip has been revolutionised, and in more recent production installations every operation is continuous, with the exception of the final annealing. As is well known, the reduction of the steel to the required thickness increases the hardness of the material and subsequent annealing is necessary to restore the ductility of the steel to facilitate further processes, involving drawing or forming, to which steel sheets and strip are applied. In present practice, the material, either in the form of sheared sheets or in coils, is annealed in batches, in closed boxes which possess a neutral or slightly reducing atmosphere. There are many designs of furnaces based on this principle, some of which are specially designed for annealing sheets, while others are concerned with sheet or strip in the form of coils.

Continuous annealing has for long been successfully used for wire, and for non-ferrous metals and alloy strip, but there were thought to be excellent metallurgical reasons against the continuous annealing of steel sheets and strip; there were also considerable difficulties connected with handling the hot strip in the furnace. For these reasons, although the cost of batch annealing and its ancillary operations is considerable, this type of treatment was felt to be indispensable for producing the required characteristics in the finished material, and the application of batch annealing is almost universal. According to Keller,\* however, recent trials of continuous furnaces have shown that continuous annealing can produce strip material having characteristics equal to or superior to those of batch-annealed material for many uses, and means of overcoming the difficulties in handling the hot strip have been developed. It is considered that with the progress of this development the production of strip will soon be fully continuous and will result in a considerable reduction in the cost of annealing, an increase in production capacity, and in practically absolute uniformity of material produced.

Attempts to introduce continuous annealing have covered a considerable period. As far back as 1915, gas-fired "blue-annealing" furnaces were used, through which the sheets were passed over rollers in the hearth. More recently a continuous furnace for normalising motor-car body sheets was installed at a large plant.<sup>1</sup> This furnace is 300 ft. long, is electrically heated, and operates at about 950° C. Within the last two years, three other strictly continuous furnaces have been installed for handling strip in coils rather than in separate sheets. One of these is called a "catenary" furnace, because the strip is suspended under some tension between two rollers at either end of the horizontal furnace, forming a catenary curve; the furnace is gas-fired, using radiant tubes, but a protective atmosphere is not used; it therefore acts as a blue-annealing rather than a bright-annealing unit. A second installation is believed to consist of a tower-type furnace, about 60 ft. high, electrically heated, through which the strip passes at a rather slow speed, somewhat over 50 ft. per min., being heated to about 940° C. and then cooled to about 65° C., both in a protective gas atmosphere. The purpose here,

however, is not annealing, but treatment of a tin layer which has previously been formed on the surface. A third installation consists of two tower-type electrically heated furnaces about 30 ft. high, each having a capacity of about 3 tons per hour, producing bright annealed strip at a speed of about 200 ft. per min. More recently, this has been increased to 320 ft. per min., and men of experience state that speeds of 500 ft. per min. and over are not impossible. The Rohn continuous furnaces, similar to the latter, are reported to be in use in Germany for bright annealing. Continuous furnaces of the horizontal type must be long if a reasonably high production rate is to be attained; the length, in fact, may be prohibitive with regard to the tension produced in the strip, while the surface of the strip appears likely to be damaged; because of these difficulties, the tower-type of furnace is considered to be preferable.

In present-day strip-mill practice, batch annealing is preceded by scrubbing and electrolytic cleaning of the strip, in order to remove the oil which is left on its surface, and, in fact, in its pores, by the pressure of the rolls in the cold-reduction mill through which it is passed. If the oil were left on the strip during the process of annealing it would cause localised discolouration of the surface, or would deposit a thin but adherent layer of carbon. Following annealing, the strip is usually again passed through the mill rolls, set to reduce the thickness very slightly, in order to temper the dead soft material. This operation reduces creasing or buckling that would otherwise occur in subsequent operations, even in coiling or uncoiling of the strip; while in drawing and forming stretcher strains would occur which disfigure the appearance of the surface. In making a comparison between continuous annealing and present annealing practice, it is necessary to consider the cost of the operation of annealing in itself, but also of the preceding cleaning and the succeeding temper-passing, both of which are expensive operations.

On the basis of the successful results obtained with electrically heated tower-type furnaces, a design of gas-fired tower-type strip annealer has been developed and is now being introduced in the steel industry. A complete full-size unit or "line," including handling equipment, is given by Mr. Keller, and is reproduced in Fig. 1, while the furnace itself, including the cooling chutes, is illustrated in Fig. 2 on a larger scale. The equipment shown in these illustrations is designed to handle steel strip from 0.0086 in. to 0.025 in. in thickness, up to 38 in. wide, at speeds of 75 ft. to 300 ft. per min. When handling the usual size—viz., strip 0.01 in. thick and 30 in. wide—at a live speed of 200 ft. per min., the output should be approximately 1,200 lb. per hour.

## Furnace and Cooling Zone

The vertical furnace, shown in Fig. 2, is about 30 ft. high, having two passes, each heated by horizontal gas-fired radiant tubes on both sides of the strip. The tubes are of U or hairpin shape, and alternate tubes along the height are entered from opposite sides of the furnace, in order to give uniformity of temperature across the width of the strip. The furnace is designed to operate normally at a temperature of 700° C., with a maximum operating temperature of 925° C.

\* Presented at the Fourth National Defence Meeting, Pittsburg, Pa., of the American Society of Mechanical Engineers, and published in *Mechanical Engineering*, vol. 63, No. 7, pp. 507-512.

<sup>1</sup> "Annealing and Normalising Auto-Body Steel in Electric Furnaces." By E. T. Hague and P. H. Brace. *Iron and Steel Engineer*, vol. 13, pp. 47-58.

The strip, on leaving the furnace, passes through six vertical cooling chutes, in which its temperature is reduced to 200° F. or lower. The gas-tight chutes are water-jacketed. In addition, the strip, in passing through them, is cooled by forced convection. Both the heating and the cooling are ordinarily effected with the strip immersed in a protective gas atmosphere, in order to retain the bright surface of the strip.

Control of the temperature in the furnace is completely automatic, with nine control zones. Each zone has a potentiometer-pyrometer controller of the proportioning

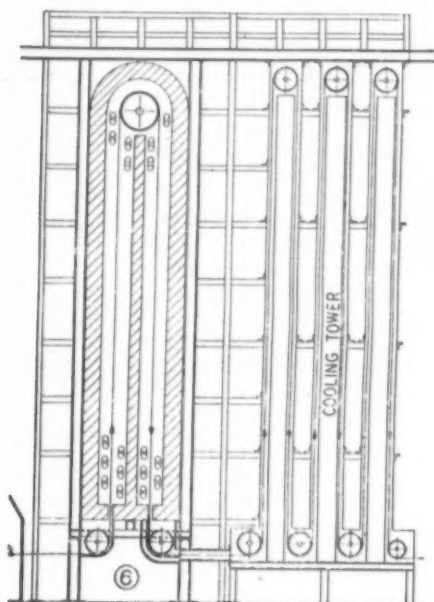
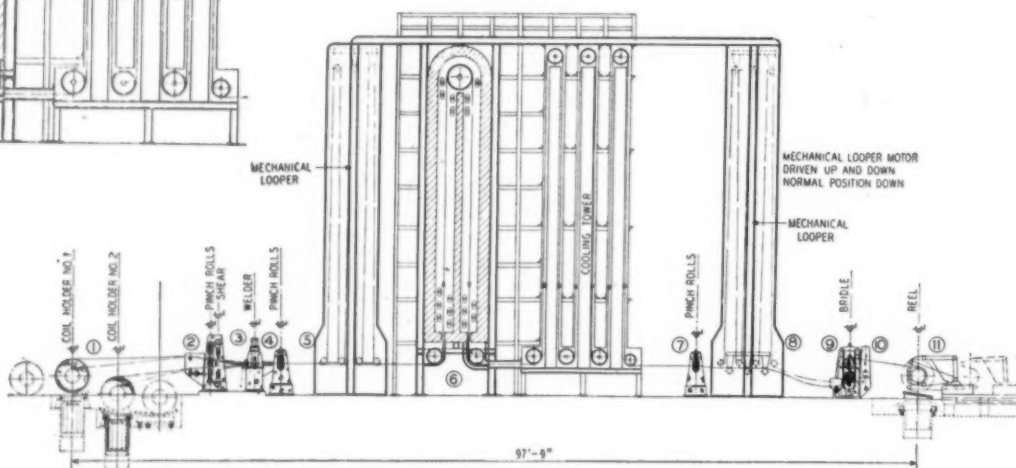


Fig. 2.—The furnace of the strip annealer, with cooling chutes.

Fig. 1.—Gas-fired tower-type steel strip annealing furnace, including handling equipment.



type, which varies the gas through throttling-type valves in accordance with the heat input required for that zone of the furnace. For the first eight zones, the potentiometer controller is of the indicating type, while for the ninth zone it is of the recording type.

#### Sequence of Operations

Assuming that the line is in operation with the strip being paid off from the No. 2 coil-holder, a coil is then entered on the No. 1 coil-holder and the end of the strip is fed by hand to the combination unit (2), where an air cylinder is operated to clamp the strip between the upper and centre pinch rolls. The leading end of the strip is then cropped by the upper shear blade, whereupon the upper pinch roll feeds the end of the strip up to the welder. Meanwhile, the strip from No. 2 coil is passing at full speed through the open lower portion of the shear-and-pinch-roll unit.

When the end of the coil being paid off from the No. 2 coil-holder is about to be reached, the operator speeds up the feeding pinch rolls (4). The counterweighted top rolls of the looper move upwards, pulling the strip with them, and accumulating approximately 120 ft. of strip in the

looper when they have reached the top of the travel. When the tail end of the coil reaches the shear (2) the feeding pinch rolls (5) are operated to bring the end of the coil into welding position and hold it there, while it is being welded to the leading end of the strip from No. 1 coil-holder. When the end is stopped, the body of the strip is being pulled through the furnace at constant rate by the pulling pinch rolls (7), being supplied from the accumulated slack in the looper. As soon as the weld is completed, the upper pinch roll of the combination unit (2) is raised free of the strip, the feeding pinch rolls (4) are started, and normal operation is resumed. It will be evident that the design provides for exceedingly fast coil entry.

After the strip has passed through the furnace and cooler, at such time as it is necessary to cut out a weld and thus separate the ends of two coils which had been joined, the reel and the tension bridle are stopped and the weld is cut out by operating shear (10). Meanwhile the pulling pinch rolls (7) draw the strip through the furnace and cooler at the normal speed, the slack being taken up by the rising of the counterweighted rolls of the mechanical looper (8).

The tail end of the sheared strip is quickly wound on the reel (11), the latter is collapsed, and the coil is quickly stripped from the reel-head by means of the coil-car and the hydraulic lift. As soon as the retiring coil clears the end of the reel, the belt wrapper is moved into position round the drum, and, by means of the tension bridle, the sheared leading end of the strip is fed up to the reel, where it is quickly and automatically entered by means of the belt wrapper. The slack in the looper (8) is taken up by

increasing slightly the speed of coiling, and normal operation then proceeds at constant rate.

#### Cleaning, Tempering and Surface Characteristics of Strip Produced

In the electrically heated tower-type continuous annealer previously mentioned it was found that no cleaning of the strip was required either before it entered the furnace or after it left. The strip coming out of the furnace and cooler unit was clean and bright, and could have been tinned directly as it came from the annealer. Actually, it was pickled before tinning. The results obtained by cleaning in the furnace and eliminating previous electrolytic cleaning were studied, and it was found that the porosity of the tin coating was approximately the same as that of regular coke tinplate of equal coating.

It was found also that the repeated reversed bending over rolls of steel strip at somewhat elevated temperatures results in a definite "tempering" of the strip material, equivalent to mild skin passing or temper rolling. Thus the temper-tower annealer for some kinds of strip product eliminates not only the slow box annealing, but also the expensive electrolytic cleaning and some part of the yet

more costly temper passing. Instead of giving the material three or four passes through a skin-pass mill, as is done in the case of box-annealed material in coils, the continuously annealed material need be given only one or two passes. Other expedients now used in connection with skin passing, such as roughening the cold-mill rolls in the first pass, can be eliminated, since the proper surface condition can be secured in the material by atmospheric treatment in the furnace.

By suitable regulation of the atmosphere, a range of types and colours of surface from that representing bright annealing to a Russian-blue surface, or any intermediate colour, can be produced.

### Physical Characteristics of Strip Material Produced

When the development of a continuous strip annealer was first considered, the opinions of metallurgists were sought as to the properties which might be expected in strip material annealed in this manner with a comparatively short heat and cooling cycle. Competent opinion at that time indicated that once the temperature had been reached the changes which result in annealing would take place almost instantly, and the same characteristics would be obtained as in box annealing. The required time of the cycle would therefore be merely that consumed in heating the strip to about 700° C. and cooling it down again. Considering the internal stresses and having in mind Maxwell's law of relaxation of such stresses, the author thought that other factors might considerably modify this conclusion. Later experiences have shown that continuously annealed short-cycle strip material in fact does not have exactly the same physical properties and grain structure as long-cycle box-annealed material. However, it has properties which for many uses are equally good and for some purposes are superior to those of box-annealed strip.

More than a year ago, tests of tensile properties, hardness, and ductility were made on non-tinned material, designated "black-plate," and on tinned material both produced by the short-cycle continuous annealing process. The tinplate had higher carbon (0.10) and high manganese (0.49) content than the "black" material, and was somewhat thinner, having a thickness of 0.01 in. The test samples were taken from the mid-section of the sheets, and the results are given in Tables I and II, where L and T indicate longitudinal and transverse tensile specimens respectively.

TABLE I.  
TENSILE PROPERTIES AND HARDNESS.

Material	Specimen	Elongation % in 2 in.	Lower Yield Strength, psi.	Ultimate Strength, psi.	Hardness, Rockwell 30T.
Black plate .....	L	29.3	39,700	55,050	58
	T	27.8	39,000	53,100	58
Tin plate .....	L	17.8	58,800	61,150	66
	T	24.8	60,000	61,700	—

TABLE II.  
OLSEN DUCTILITY TESTS.  
(Average of three tests).

Material	Deformation,	
	Pressure, Lb.	In.
Black plate .....	1,025	0.298
Tin plate .....	934	0.282

In general, the material tested appeared to be fairly stiff, which is advantageous provided it does not mean sacrifice of drawability; and the good elongation results indicated that the drawability should be adequate for many purposes. It was feared, however, that uneven forming might result from the stiffness, and that stretcher strains might develop, due to the fine grain structure. Since that time, however, it has been possible to reduce the hardness. At present, material down to Rockwell 48 is being made by continuous annealing. Indications are that it will be possible to go as low as Rockwell 44. But it has been found that, because of the difference in grain structure, the box-annealed and the continuous annealed material cannot be compared exactly on the basis of their Rockwell hardness values.

A material of higher Rockwell hardness from a continuous annealer has been found to behave exactly as well in forming operations as box-annealed material of much lower Rockwell number, although some changes in the forming practice may be required. In the plant referred to all but two classes of strip material are continuously annealed; the exceptions being very thin material (0.0065 in. thick) which is subject to breakage, and material which is to be drawn to form very deep containers, and which must therefore be dead soft.

It is evident, therefore, that the process of continuous strip annealing has reached a state of development at which tinplate and similar thin material can be produced having characteristics which make it suitable for many uses. Furthermore, development work is continuously in progress which promises to result in a still further improved product.

### The X-Ray Spectroscopic Method

AN apparatus known as the X-ray spectrograph has, we understand, been designed and constructed by Messrs. I. Borovosky and W. Blockin, of the X-ray chemical laboratory at the Geological Institute of the Academy of Sciences of the U.S.S.R. With this apparatus it is claimed to be possible to detect quantities as minute as one five-millionth of a gramme of a rare element in a hundredth of a gramme of substance.

The first of these X-ray spectrographs has been installed in the universities and research institutes in Moscow, Tbilisi, Eriven, Kiev, Odessa and Sverdlovsk, and special laboratories employing the X-ray spectroscopic method for analysing ores and alloys are being set up in the important centres of the U.S.S.R.

### Basic Refractories

In the ferrous and non-ferrous metal industries refractories play a vital part in the present need for increased production, and of late increasing attention has been directed to this aspect of greater outputs. For many years the highly refractory qualities of magnesite, chrome, and more recently of chrome-magnesite, have been known, and such refractories are now widely used in their industries. It is only within recent years, however, that improvements have been made in manufacture, giving refractories which are capable of withstanding the severest operating conditions.

In no branch of refractories have such important advances been made as in chrome and basic refractories. The main advantages arising from their adoption are higher heats, faster driving, greater outputs and longer campaigns; the wise use of these refractories can effect substantial economies in operating costs and can give greatly increased outputs. In this connection it is noteworthy that John G. Stein & Co. Ltd., by co-operating with two of the pioneers in this field, have been able to install modern plant for producing these refractories of the highest quality. Their research and development facilities enable the firm to approach nearer to the stringent requirements of modern furnace practice.

Six brands of basic and chrome refractories, each possessing its own particular properties and covering the needs of the ferrous and non-ferrous metal industries, are discussed in an elaborately produced booklet which gives the characteristics of each brand as well as a list of applications. The booklet which contains some value data and excellent sectional illustrations of furnace plant, is available on application to John G. Stein & Co. Ltd., Bonnybridge, Scotland.

THE Carborundum Company Ltd., Trafford Park, Manchester, inform us that their London Office address is now : 3 and 4, Parkfield Parade, High Road, Ickenham, Middlesex. Telephone : Ruislip 3586.



## The Heat-Treatment of Aluminium and its Alloys

**T**HE growing use of the heat-treatable alloys of aluminium in the various industries has resulted in an increasing demand for information concerning heat-treatment technique and practice adopted with these important materials. In an effort to meet this demand the Research and Development Department of the Northern Aluminium Co. Ltd. has devoted its latest bulletin to a comprehensive survey of modern aspects of the heat-treatment of these alloys.

It is known that certain constituents of aluminium alloys, notably the copper-aluminium compound  $\text{CuAl}_2$ , the magnesium-silicon compound  $\text{MgSi}$ , and also magnesium and zinc, are soluble in aluminium to a much greater extent at high temperatures (approaching the alloy's melting point), than at low temperatures. If the alloy is at high temperature for a period sufficiently long for the solid solution of these constituents to take place, and is then rapidly cooled by quenching in water, an artificially high amount of constituent remains dissolved throughout its mass. This solid solution is not stable at room temperature, and the result is that the compound or element in excess tends to be precipitated in sub-microscopic form. The precipitation of these compounds appears to have a "keying" action between internal slip planes, helping to prevent or regulate motion and thereby increase the mechanical strength and in some cases the ductility of the material.

The extent of this natural precipitation or ageing, and the temperature at which it will take place, can be determined to a considerable extent by adjustment of the chemical composition, and in the wrought class two main types of heat-treatable alloy have been developed:—

- (a) The single heat-treated or naturally ageing alloys. These are given a high temperature treatment followed by a quench in water, and their strength then rises spontaneously without further heating. After about five days at room temperature their strength reaches the maximum value. This process is termed solution heat-treatment.
- (b) The double heat-treated or artificially aged alloys which are only hardened to a limited amount by solution heat-treatment and quenching, but which develop their maximum strength properties by means of a subsequent precipitation treatment which is effected at a lower temperature than for solution treatment.

The extent to which the increase of strength and hardness, with respect to time, influences forming practice adopted with these types of alloys is reviewed from an essentially practical angle.

A considerable part of the bulletin is naturally devoted to the subject of solution treatment practice, particular stress being laid upon the necessity for the strict observance of correct temperatures and the allowance of sufficient time at the treatment temperature to ensure complete solution of the hardening constituents. In this latter connection it is pointed out that the time of soaking will depend upon the physical condition of the metal, the size of the article to be treated and the type of furnace used.

For a given size, castings require the longest soaking times, since most of the hardening constituents are of coarse structure and consequently take longer to dissolve. Due to the relative softness of aluminium alloys at solution treatment temperatures, large castings require careful handling, and may require support to avoid distortion during treatment. In wrought products, such as sheets, forgings and extruded sections, the working to which the material has been subjected causes breakdown and dis-

persal of the hardening constituents, and they are, therefore, more readily soluble.

The heavier the gauge of the stock to be treated the longer will be the time of soaking, but it is pointed out that, since the treatment time will also depend on the type of furnace used, it is unwise to make definite recommendations in this respect.

Deep pressing work and the manipulation of intricate components are often found to necessitate repetitions of solution treatment before the required degree of deformation is achieved. Careful control of conditions permits of a reasonable number of reheat-treatments without harmful effect on the metal, but in the case of "Alclad" sheet, particularly of light gauge, there is a tendency for the alloying elements of the core to diffuse into the high purity coating. This diffusion, if excessive, will cause some reduction in resistance to corrosion and, consequently, it is advisable that the number of solution heat-treatments applied to "Alclad" be kept at a minimum.

The choice of heating equipment is influenced by the form of product being treated, but the fused salt bath, heated by gas, oil or electricity, is the most widely used type due to its high heat capacity and the fact that, with reasonable care, uniformity of temperature is easy to maintain. Results obtained with high-speed air furnaces have been found to be satisfactory providing that attention is given to efficient spacing and furnace atmosphere, and this type of equipment has been increasingly used during recent years. The precautions necessary to ensure continued accurate operation of heat-treatment equipment are detailed in the bulletin and treatment temperatures for both wrought and casting alloys specified. Also tabulated are times of immersion in the salt bath for various gauges of sheet and strip and sizes of rod, bar and extruded sections, but it is pointed out that the times given are intended as a guide only and that local conditions will generally involve some modification. Quenching problems and the removal of adhering nitrate from the metal are discussed, and the causes and results of heat-treatment troubles are fully dealt with and illustrated. Refrigeration storage, to extend considerably the period following solution treatment during which age hardening alloys may be subjected to cold forming, is an interesting feature. The cold storage of sheet, in addition to the older practice as applied to rivets, is explained and interesting examples illustrated of refrigerator mobile equipment for the conveyance of blanks from the refrigerated store to the presses.

Precipitation heat-treatment or, as it may alternatively be referred to, artificial ageing practice is reviewed, chiefly from the point of view of the attainment of optimum mechanical properties. In both cases descriptions and explanations are supported by tabulated data.

The opportunity has also been taken to clarify problems relating to annealing, particularly those concerned with the "strong" aluminium alloys. To this latter end, discussion of the changes involved, from the points of view both of physical structure and mechanical properties, is dealt with in sections devoted to strain hardening wrought alloys, heat-treatable alloys which have been strain hardened only, and the fully heat-treated alloys. In this way it has been found possible to deal concisely with the removal of hardening resulting from cold work and/or heat-treatment.

The final sections are devoted to the pre-heating of forging stock and the care of pyrometric equipment, and emphasise the essential practical techniques which should be followed if trouble is to be avoided in both the working and treatment of aluminium alloys.

Copies of the bulletin may be obtained by applying, on business notepaper, to the Research and Development Department, Northern Aluminium Company Ltd., Banbury, Oxon.

# Chemical Black Finishing of Zinc Base Alloy Castings

By "Zincus"

*With the increasing use of zinc-base alloy components, particularly in the form of die-castings, more attention is being directed to their finish. In many cases an inexpensive chemical black finish is of value in enhancing the appearance of the finished product, and several types of solutions are discussed which give this finish.*

**I**N the present national emergency it has become necessary in many cases to substitute zinc base alloy pressure and gravity die-castings in applications where light alloys were previously used, in order that aluminium may be diverted to more important functions. In many applications of zinc base alloys, the provision of an inexpensive chemical black finish is of value, not only in enhancing the appearance of the finished product, but also in saving the materials and time consumed in the more elaborate procedures required for spray painting, enamelling and plating. It is proposed to review, therefore, the various methods that can be adopted for the black finishing of zinc base alloy products by simple immersion processes, and to indicate from practical experience the technique necessary to ensure good results.

The first essential to secure good results is, with zinc as with every other metal, efficient pre-treatment to ensure that the surface is free from adherent grease and dirt—in other words, the metal surface must be chemically clean. If a slight surface roughening is not objected to in the finished product, the easiest and at the same time one of the most efficient methods of pre-treatment is to subject the castings to a light sandblasting, which not only removes grease and dirt, but leaves the surface in the most reactive condition for the subsequent blacking operation. It must be borne in mind when considering sandblasting, however, that the resulting black finish will invariably be more matt than if the chemical treatment is carried out on the "as cast," polished, or machined surface. The sandblasted castings should not be handled with the bare hands, as the surface will readily stain and absorb grease from the skin. Handling from the sandblast booth to the immersion blacking tank should be carried out either with tongs or a gloved hand.

If a glossy black finish is desired, it is essential that the metal should first be polished, this operation being carried out according to normal works' practice. Polishing grease should first be removed, either in a cold bath of solvent, such as naphtha, xylol, or the like, but preferably in cold trichlorethylene, which is non-inflammable, or even better, in a vapour-phase plant of the well-known I.C.I. degreaser type, using trichlorethylene as the vaporised solvent. Castings in the "as cast" condition, which have become coated with oil or grease, should also be subject to solvent treatment.

The final removal of all traces of oil and grease from "as cast" and polished castings is then conducted, using a hot mild alkaline solution followed by a weak acid dip. Several types of solutions are suitable for the hot alkaline cleaner and some typical formulae are given below:—

- (a) Immersion in a solution containing 2 oz. trisodium phosphate and  $\frac{1}{4}$  oz. sodium hydroxide (caustic soda) per gal. for 1 to 2 mins. at 70°–80°C.
- (b) Immersion for 10 to 15 mins. in a boiling solution containing 6 oz. trisodium phosphate per gal.
- (c) Immersion at 90°–95°C. in a solution containing 1 part of saturated solution of sodium carbonate in water to 1 part of a 6% solution of sodium silicate in water.

All of the above solutions may be used in ordinary mild

steel tanks heated with gas burners, and whichever solution is used the mild alkaline cleaning operation should always be followed by thorough rinsing in cold running water. The rinsing operation will not, however, completely remove the thin film of alkaline zinc salts left on the surface of the work from the cleaner, so that the articles should next be immersed for a few seconds in a cold 10% solution of hydrochloric acid contained in an earthenware vat.

The period of immersion should, in general, be not more than for a few seconds until an even evolution of gas is seen to take place over the entire surface of the articles when they should be withdrawn and again thoroughly rinsed in cold running water. This acid etch operation is of particular importance in that it greatly facilitates the production of an even black finish in the subsequent chemical immersion process.

It is obvious that articles in as "as cast" condition will require a longer period of etching than goods which have been polished in order to break up slightly the hard "as cast" skin, whereas polished castings require only slight etching, otherwise the gloss of the finish will be diminished. The precise degree of etching necessary can, however, quickly be determined after a few practical trials.

There are several methods available for the production of a black finish by simple immersion, but whichever is used the preliminary cleaning operations remain the same.

## Immersion Processes

To produce a black colour on zinc it is necessary to form on the surface of the metal an adherent coating of a metallic compound other than zinc itself, since most zinc compounds are white. Many of the older methods depend on the formation on the zinc of a coating of black copper oxide, but more recent processes aim at the production of compounds of nickel, manganese and molybdenum. Some of the methods advocated are laborious and not too reliable in ultimate result. It is not proposed, therefore, to describe these in detail, but to refer only to processes which from the point of view of ease of application and consistency of finish can be relied upon to give satisfactory working.

Dealing first with one of the best of the solutions for colouring zinc black by the formation of a copper compound; this is made up by dissolving in 1 gal. of water  $3\frac{1}{4}$  oz. each of (a) copper nitrate, (b) ammonium chloride, and (c) concentrated hydrochloric acid.

This solution is used cold and is best held in a stoneware vat, enamelled bowl or rubber lined steel tank, since if bare metallic tanks are used the copper salt will be wasted by precipitation of metallic copper. After the articles to be blacked have been chemically cleaned in the manner previously described they are dipped into the above solution at room temperature for about ten seconds, withdrawn, and at once swilled well in cold water. Hooks made from galvanised iron wire are very convenient for handling the parts into the immersion and swill tanks.

After well swilling, the blackened articles are hung in boiling water until they have acquired sufficient heat to dry of their own accord when withdrawn from the tank. The parts will then be found to be covered with a loose

black "smut" under which is a tightly adherent black coating. This "smut" can be removed by any convenient means, but perhaps the best way is by brushing with a stiff bristle brush lightly greased with lanoline.

The operative should be provided with a range of differently shaped and sized brushes for dealing quickly with awkwardly shaped recesses, when it will be found that with a little experience the loose surface deposition can be removed in a few seconds to leave the article with an ebony black and slightly glossy finish. The thin coating of grease that results from the brushing operation is beneficial, as it not only deepens and enhances the tone of the black produced but also forms a protection against weathering effects. If quantity production is desired, the provision of bristle brushes revolving on standard polishing lathes is a time-saving refinement.

In some cases it may be found that, with certain types of zinc base alloys, the solution described gives a black with a reddish tinge. If this occurs the reddish tinge can be suppressed by adding to the solution a small quantity (which can quickly be determined after a few trials) of antimony chloride.

The following solution, of more recent origin, is rather slower in working than the one previously described, but has the advantage that there is little or no formation of black "smut," so that the brushing operation is no longer essential.

Ammonium Chloride	..	..	4 oz.
Nickel Chloride	..	..	6 oz.
Ammonium Sulphocyanide	..	..	2 oz.
Zinc Chloride	..	..	$\frac{1}{2}$ oz.
Water	..	..	1 gal.

This solution is best held in a water-jacketed earthenware vat or in a vitreous enamelled bowl heated in both cases by gas. The previously degreased and etched parts are immersed in the solution at 100°-120° F. for a sufficient time to become completely blackened, which usually takes from 1 to 3 mins. As before, they are then swilled well in cold water, and held in hot water for a short time and finally allowed to dry in air. Although there is little or no "smut" formed by this solution, a quick brushing of the exterior of the casting with a dry bristle brush will be found to improve the sheen of the finish.

This solution, which actually deposits a coating of "black nickel" on the zinc, is especially to be recommended for the treatment of polished articles, the resulting finish being a glossy black of excellent appearance. It is of importance that a bath of this type produces a consistent black finish with no trace of the reddish tinge which may sometimes arise from solutions containing copper salts.

Morgan and Lodder, at a meeting of the Electro-depositors' Technical Society, described a process for the production of a black finish depending on the known properties of certain organic zinc salts to absorb dyestuffs in a manner similar to that of the aluminium oxide film on anodised aluminium. Although the capacity of such organic zinc salts to absorb dyestuffs is relatively feeble when compared with the anodic film on aluminium, it is nevertheless possible under controlled conditions to produce on the zinc articles a relatively impermeable coating which can be dyed black in a suitable solution and is of good wear resistance. The bath for producing the absorbent organic salt coating is prepared by dissolving 1 lb. of oxalic acid crystals in 1 gal. of hot water, held in an earthenware tank.

This solution is allowed to cool down to room temperature before use to produce a saturated solution of oxalic acid in which the previously cleaned and etched castings are immersed for 10 mins. At the initial stages of immersion there is a rapid evolution of gas, but this becomes negligible after two to three minutes. The castings are afterwards swilled well in clean cold water, care being taken not to damage the coating by knocking castings together, as at this stage it is relatively fragile.

The swilled castings are then immersed in a black dye solution at 90° C. containing  $\frac{1}{2}\%$  of "Black I.N. dye" for

one hour. At the end of this time they are well rinsed in running water and allowed to dry off, after which they are lightly rubbed with a soft rag to remove any traces of "bloom" from the surface. It is claimed that 230 sq. ft. of casting area can be processed for the consumption of 2 oz. of oxalic acid. The black dye used is apparently a special one which can be obtained from the Imperial Smelting Corporation Ltd. The finished coating is of excellent appearance and has good resistance to general handling.

Excellent black coatings can be produced from solutions containing ammonium molybdate and addition agents, and much research has been and is being carried out on processes of this type. Molybdate finishes are fairly tough and wear-resisting, and no "smut" is produced during the coating operation.

The trend of development in this direction is well indicated by U.S. Patent No. 2,219,977, granted to O. Brill, assignor to Mannesmann Stahlblechbau A.G., and dated October 29th, 1940. This patent covers a method for colouring zinc black, in which the metal is first dipped in a solution of a heavy metal sulphate to deposit a metallic film by immersion, and then in a buffered molybdate solution at a pH value of about 5.

As a typical example, zinc can be dipped for about 60 secs. in a solution held at room temperature and containing 10% nickel sulphate and 1% acetic acid, or 10-15% copper sulphate with 1% of acetic acid. If this second solution is used the dipping period is only 30 secs. After swilling, the parts are then treated for 3-6 mins. in a solution at 50°-60° C. containing  $\frac{1}{2}\%$  to 5% of ammonium molybdate and  $\frac{1}{10}\%$  to  $\frac{1}{2}\%$  of tartaric acid.

There are, however, more simple ways of producing black colours from molybdate solutions. According to J. B. Kushner, excellent results can be obtained by dipping the previously cleaned zinc parts for a few seconds at 180° F. in a solution containing 2 oz. ammonium molybdate, 4 oz. ammonium chloride, 1 oz. potassium nitrate, and 1 oz. boric acid per gal. of water.

Other useful solutions, all of which will produce good finishes on zinc, contain per gal. :-

- (a) 3 oz. ammonium molybdate +  $1\frac{1}{2}$  oz. ammonium chloride.
- (b) 3 oz. ammonium molybdate + 1 oz. sodium acetate.
- (c) 3 oz. ammonium molybdate +  $\frac{3}{4}$  oz. oxalic acid.

Solutions (a) and (b) are best used hot and (c) cold, but in all cases should be held in earthenware vats or enamelled bowls.

Good black finishes can be produced by any of the foregoing processes described, but it should be remembered that success will largely depend on the efficiency with which the preliminary degreasing and etching processes are performed. Improved resistance to handling and weathering is achieved if the treated articles are clear lacquered or slightly greased.

If a thin grease coating is desired, this can be quickly and easily given by dipping the castings in a solution having a small amount of lanoline dissolved in solvent naphtha or white spirit. The solvent volatilises to leave a thin and evenly disposed coating of grease on all surfaces.

Clear lacquering is advantageous wherever it can be applied, as it provides a protective coating for the finish, while being practically unnoticeable both to touch and feel.

### Personal

Colonel W. C. Devereux has been appointed Controller of North American Aircraft Supply in the Ministry of Aircraft Production. He has resigned, for the period that he holds this office, from his directorships of all the companies with which he was associated that are engaged on the production of aircraft materials. Colonel Devereux has also been elected a Vice-President of the Royal Aeronautical Society for the year October, 1941, to September, 1942.



# The Application of Spectrographic Methods to the Analysis of Segregates\*

By F. G. Barker, F.I.C., J. Convey, Ph.D., and J. H. Oldfield

*Experiments have been carried out by the spark method of spectrographic analysis with a view to the development of a suitable technique for the analysis of segregates. As a result of tests conditions are given under which the elements—silicon, manganese, nickel, chromium and molybdenum can be estimated from one exposure of 15 sec. It is shown that tests on small bars of nickel-chromium-molybdenum steel give reproducibility of a satisfactory order, and that tests on specimens of large forgings give higher average values for all elements in the segregates than in the normal metal. This is an abridgement of the original paper*

THE possibility of obtaining at least a qualitative estimate of the composition of metallic segregates was shown in an earlier paper by Barker.<sup>1</sup> At that time, however, no systematic study of the problem had been attempted, and the work done was chiefly to show the extent to which the presence of segregates might influence the results of routine spectrographic analysis, rather than to determine their actual composition. The segregates examined were large, and the technique employed was the same as that used for the routine spectrographic analysis of steel. The matter aroused some interest, and the author was approached by the Chairman of the Inclusions Sub-Committee of the Committee of the Heterogeneity of Steel Ingots to pursue the matter further, and work out, if found practicable, a technique for the quantitative analysis of segregates. This was undertaken, and the results of the work are recorded in the present paper.

From the results of early experiments it soon became evident that, if the method was to be applicable to any but large segregates, the technique employed for routine analysis would have to be modified, so as to reduce the area affected by sparking. It was thought that this could be effected by a change in the shape of the electrode point and by the use of a material which would keep its point better than graphite. Silver appeared to be the most suitable metal, and a series of tests were made with electrodes of this metal. Tests were also made to determine the extent to which the sparking area could be reduced by a reduction in the spark gap. Tests were made with gaps of various amounts down to 0.002 in. With a decrease in gap there was a decrease in the sparking area, and the best results with regard to uniformity of crater, shape and halo were obtained with a nipple electrode. All subsequent work was, therefore, done with electrodes of this type. With the smallest gap tested, —viz., 0.002 in. and a 15-sec. exposure, the diameters of the crater and halo formed with this electrode were 0.008 in. (0.2 mm.) and 0.088 in. (2.2 mm.), respectively. As a result of these and further tests, the following conditions were provisionally adopted for the analysis of segregates:—

Wave-band .....	2,200—3,000 Å.
Slit width .....	0.0006 in.
Distance of electrode from slit .....	4 in.
Spark gap .....	0.015 in.
Voltage .....	15,000.
Added inductance .....	Nil.
Upper electrode .....	Silver rod, nipple-pointed and oxidised by pre-sparking for 15 secs.
Plate .....	Kodak Photoresist B10.
Exposure time .....	15 secs.

Sparking under the above conditions, a polished specimen gave a diameter of the crater of approximately 0.012 in.

and that of the halo approximately 0.10 in. The depth of the crater as measured with a Société Genevoise measuring instrument, type *M U-214B*, on which an accuracy of  $\pm 0.00005$  in. is claimed, is 0.0007 in. The weight of steel corresponding to the volume of the crater is approximately 0.000005 g.

Standard curves, relating the log. ratio of the line intensity to the percentage of element present, which were prepared from chemically-standardised steels, are shown in Figs. 9 to 13; each point on the curves is the average of ten spectrographic determinations.

Before commencing the examination of segregates it was necessary to establish whether the results of tests were dependent entirely on the spectral emission resulting from the main central discharge which forms the crater, or whether they were influenced by emission from the zone around the crater which has the appearance of a halo. The following tests were therefore made. A  $\frac{5}{8}$ -in. hole was drilled and ground in the centre of a piece of mild steel  $1\frac{1}{4}$  in. square by 1 in. thick, which did not contain more than traces of nickel, chromium and molybdenum. A piece of nickel-chromium-molybdenum steel was ground slightly taper to make a tight fit in the mild-steel block, and was forced in under a hydraulic ram. The face was ground, polished and very lightly etched. A number of spectrograms were then prepared by sparking on the mild steel in such positions that the resulting halo overlapped the alloy steel by various amounts. An examination of the spectrograms showed that not more than faint traces of the alloy elements appeared until the edge of the crater was at the actual junction of the two metals, thus establishing that emission from the halo does not influence the results of tests.

An examination of a number of segregates occurring in large forgings of nickel-chromium-molybdenum and nickel-vanadium steels was next undertaken. Specimens of the former were obtained from a large hollow cylindrical forging. Most of the segregates were roughly circular in cross-section, having a diameter of approximately 0.04 in. to 0.08 in. Their length, however, was considerably greater, and on this account longitudinal sections were prepared for spectrographic examination. As the segregates were not visible on polished sections, an etch was necessary to locate them. Of the various etching reagents in common use, previous experience had indicated that Oberhoffer's reagent was one of the most suitable, because the macrostructure and the segregates were readily revealed with the minimum of attack on the specimen.

After etching, a number of separate exposures were made on the segregate and on the normal metal in positions well removed from it. The results of analysis are given in Table I.

\* Paper No. 7/1941 of the Committee on the Heterogeneity of Steel Ingots, Iron and Steel Institute, Autumn Meeting, 1941. Advance copy.  
1 F. G. Barker, *Jour. Iron and Steel Inst.*, 1939, 1, pp. 211.

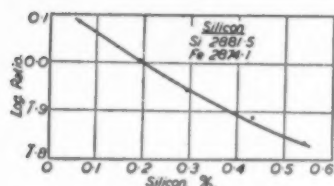


Fig. 9.—Graph for the determination of silicon from spark spectrograms.

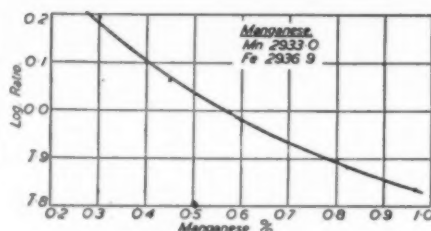


Fig. 10.—Graph for the determination of manganese from spark spectrograms.

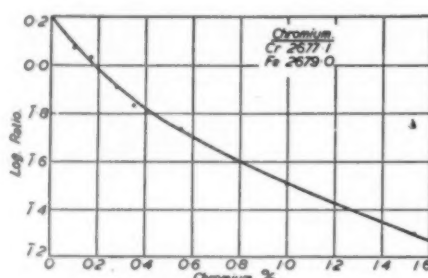


Fig. 12.—Graph for the determination of chromium from spark spectrograms.

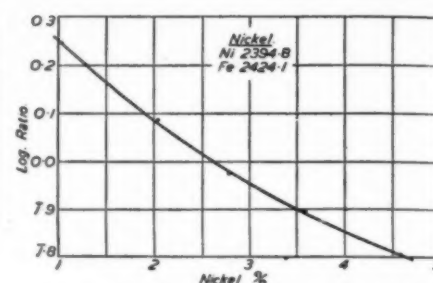


Fig. 11.—Graph for the determination of nickel from spark spectrograms.

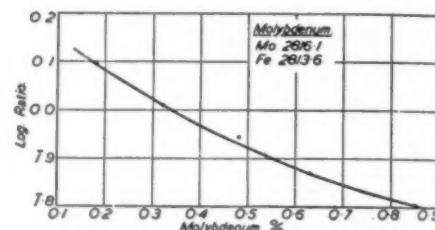


Fig. 13.—Graph for the determination of molybdenum from spark spectrograms.

Another specimen cut from a forging of similar composition which contained segregates of a similar type was also examined, with the results given in Table II. The results on the non-segregated parts of these specimens show greater variation than that generally obtained under normal conditions of routine analysis. Such variation, however, is not surprising, having in view the smallness of the area affected by sparking and the coarseness of the macrostructure. Much greater variation is shown in the results obtained from the segregates. The greatest differences occur on the manganese and molybdenum contents, the former being accounted for by the large inclusions of manganese sulphide, which were found to be much more numerous in the segregates than in the normal metal. An examination of the average compositions of the normal metal and the average of the segregates shows higher values for the latter for all elements; the greatest increases, however, occur with manganese and molybdenum, particularly the latter, where it amounts to 50% and 45% respectively, on the two forgings.

TABLE I.  
RESULTS OF TESTS ON NICKEL-CHROMIUM-MOLYBDENUM STEEL A.

Test No.	Silicon, %	Manganese, %	Nickel, %	Chromium, %	Molybdenum, %
<i>Tests on Non-Segregate.</i>					
1	0.15	0.49	2.77	0.65	0.43
2	0.17	0.46	2.68	0.59	0.42
3	0.18	0.29	2.68	0.59	0.44
4	0.17	0.46	2.55	0.55	0.39
5	0.17	0.47	2.75	0.62	0.44
6	0.18	0.46	2.68	0.59	0.43
7	0.17	0.45	2.60	0.56	0.38
8	0.16	0.49	2.65	0.56	0.39
9	0.17	0.47	2.73	0.61	0.42
10	0.17	0.46	2.68	0.58	0.40
Average .....	0.17	0.47	2.68	0.59	0.41
Greatest deviation from average ..	+0.01	+0.03	+0.09	+0.06	+0.03
	-0.02	-0.02	-0.13	-0.04	-0.03
<i>Tests on Segregate.</i>					
11	0.20	0.47	3.02	0.71	0.88
12	0.21	—	3.07	0.73	0.78
13	0.21	0.66	2.94	0.72	0.63
14	0.21	0.73	2.90	0.68	0.66
15	0.21	0.49	2.78	0.61	0.52
16	0.19	0.53	2.82	0.65	0.53
17	0.19	0.49	2.87	0.62	0.51
18	0.21	0.53	3.00	0.72	0.54
19	0.19	0.50	2.76	0.64	0.55
20	0.21	0.54	3.01	0.75	0.57
Average .....	0.20	0.55	2.92	0.68	0.62
Greatest deviation from average ..	+0.01	+0.18	+0.15	+0.07	+0.26
	-0.01	-0.08	-0.16	-0.07	-0.11

The results of tests made on segregates found in a nickel-vanadium steel forging in a similar manner to those described above are given in Table III. Here again, higher average values for all elements were obtained on the segregates, and the greatest increases are shown for manganese, molybdenum and vanadium.

The problem of assessing the accuracy of these results is made very difficult because it is not possible when testing a heterogeneous material such as steel, under conditions such as those described above, to determine to what extent variability in the results is due to actual variation in composition and to what extent it is due to experi-

TABLE II. RESULTS OF TESTS ON NICKEL-CHROMIUM-MOLYBDENUM STEEL B.					
Test No.	Silicon, %	Manganese, %	Nickel, %	Chromium, %	Molybdenum, %
<i>Tests on Non-Segregate.</i>					
1	—	0.44	2.58	0.67	0.42
2	0.16	0.45	2.62	0.65	0.39
3	0.17	0.50	2.55	0.59	0.40
4	0.17	0.49	2.63	0.66	0.44
5	0.18	0.47	2.65	0.63	0.47
6	0.19	0.49	2.62	0.60	0.44
7	0.17	0.47	2.61	0.60	0.39
8	0.16	0.44	2.55	0.59	0.39
9	0.16	0.49	2.61	0.58	0.38
10	0.16	0.48	2.70	0.66	0.44
Average .....	0.17	0.47	2.61	0.62	0.42
Greatest deviation from average ..	+0.02	+0.03	+0.09	+0.05	+0.05
	-0.01	-0.03	-0.06	-0.04	-0.04
<i>Tests on Segregate.</i>					
11	0.20	0.49	2.92	0.68	0.63
12	0.21	0.51	2.63	0.59	0.51
13	0.19	0.49	2.80	0.62	0.51
14	0.20	0.52	2.83	0.66	0.66
15	0.21	0.94	2.91	0.62	0.60
16	0.20	0.76	2.95	0.64	0.56
17	0.22	0.68	3.12	0.78	0.74
18	0.20	0.77	2.91	0.67	0.65
19	0.20	0.74	2.63	0.57	0.48
20	0.20	0.60	3.06	0.74	0.68
Average .....	0.20	0.65	2.86	0.66	0.60
Greatest deviation from average ..	+0.01	+0.29	+0.26	+0.12	+0.11
	-0.01	-0.16	-0.23	-0.09	-0.09

TABLE III. RESULTS OF TESTS ON NICKEL-VANADIUM STEEL FORGING.						
Test No.	Silicon, %	Manganese, %	Nickel, %	Chromium, %	Molybdenum, %	Vanadium, %
<i>Tests on Non-Segregate.</i>						
1	0.29	0.76	2.47	0.15	0.19	0.14
2	0.23	0.73	2.63	0.14	0.10	0.06
3	0.24	0.65	2.49	0.13	0.10	0.07
4	0.24	0.62	2.50	0.14	0.12	0.07
5	0.25	0.63	2.62	0.15	0.11	0.11
Average .....	0.25	0.68	2.54	0.14	0.12	0.09
Greatest deviation from average ..	+0.04	+0.08	+0.09	+0.01	+0.07	+0.05
	-0.02	-0.06	-0.07	-0.01	-0.02	-0.03
<i>Tests on Segregate.</i>						
1	0.26	0.73	2.80	0.15	0.13	0.16
2	0.28	0.76	2.73	0.13	0.13	0.16
3	0.24	0.82	2.75	0.17	0.19	0.19
4	0.24	0.82	2.67	0.16	0.20	0.21
5	0.27	0.85	2.73	0.16	0.20	0.18
Average .....	0.26	0.80	2.74	0.15	0.17	0.18
Greatest deviation from average ..	+0.02	+0.05	+0.06	+0.02	+0.03	+0.03
	-0.02	-0.07	-0.07	-0.02	-0.04	-0.02

\*Approximately.

mental error in the technique employed. It has been shown that the weight of metal actually involved in making an analysis is approximately 0.000005 g., and, even though it were possible to isolate such an amount for chemical examination, it is quite beyond the scope of the most

modern methods of micro-analysis to obtain quantitative results from such a quantity.

It appeared that the best that could be done with the materials available was to obtain further evidence of reproducibility on bars of small section in which heterogeneity was less evident than in large forgings. Cross-sections of 1½-in. and 1-in. dia. bars were therefore prepared by polishing and etching, and ten tests made on each. The results of these tests are given in Tables IV and V.

These results are considered to indicate reproducibility of a satisfactory order, and tend to confirm that the very much wider differences experienced with forgings between the results on normal metal and those on segregates are due largely to differences in actual composition.

TABLE IV.  
RESULTS OF TESTS ON NICKEL-CHROMIUM-MOLYBDENUM STEEL BAR, DIA. 1½ IN.

Test No.	Silicon, %	Manganese, %	Nickel, %	Chromium, %	Molybdenum, %
1	0.17	0.50	3.16	0.76	0.44
2	0.18	0.52	3.09	0.77	0.42
3	0.17	0.58	3.07	0.78	0.43
4	0.17	0.55	3.15	0.77	0.44
5	0.17	0.55	3.05	0.74	0.44
6	0.18	0.52	3.16	0.71	0.43
7	0.19	0.56	3.07	0.74	0.41
8	0.19	0.56	3.05	0.72	0.42
9	0.19	0.55	3.06	0.78	0.42
10	0.18	0.55	3.08	0.77	0.42
Average .....	0.18	0.54	3.10	0.76	0.43
Greatest deviation ..	+0.01	+0.04	+0.06	+0.02	+0.01
from average ..	-0.01	-0.04	-0.05	-0.05	-0.02

A number of the spectrograms obtained in the course of the tests reported above have been examined for carbon, but the results, even on the small bars, showed such variation as to make it doubtful whether the conditions of test employed are satisfactory for this element. No results

have therefore been reported, and further work to determine the best conditions for the estimation of carbon is in hand.

It should be pointed out that, when making the tests on segregates, the actual points tested did not in many instances fall on the axis of the segregates, and their positions relative to the axis were generally irregular. It is considered that the variation in composition which is indicated to exist within a single segregate is probably due to this cause, and an effort is now to be made to determine the graduation in composition from one side to the other. It is proposed to carry out tests under conditions in which the photographic plate and the specimen are slowly traversed during sparking, their movements

TABLE V.  
RESULTS OF TESTS ON NICKEL-CHROMIUM-MOLYBDENUM STEEL BAR, DIA. 1 IN.

Test No.	Silicon, %	Manganese, %	Nickel, %	Chromium, %	Molybdenum, %
1	0.18	0.65	2.66	0.68	0.49
2	0.20	0.65	2.66	0.67	0.47
3	0.20	0.64	2.66	0.64	0.51
4	0.21	0.67	2.65	0.65	0.50
5	0.22	0.64	2.55	0.68	0.55
6	0.21	0.65	2.65	0.61	0.50
7	0.21	0.66	2.66	0.67	0.55
8	0.21	0.64	2.57	0.67	0.51
9	0.21	0.68	2.66	0.63	0.57
10	0.21	0.69	2.59	0.70	0.55
Average .....	0.20	0.66	2.63	0.67	0.52
Greatest deviation ..	+0.02	+0.03	+0.03	+0.03	+0.05
from average ..	-0.02	-0.02	-0.08	-0.01	-0.05

\* Approximately.

being synchronised, so that a continuous spectrogram showing the graduation in composition across the segregate may be obtained. Much work, however, will have to be done to establish the best conditions for such a test, and this will form the subject of a future paper.

## Induction-Hardened Cylinder Bores

**M**ETALLURGICAL engineers have been extremely fortunate in the variety and utility of hard materials and surface-hardening processes that have become available to them during recent years. Hard surfacing metals of the Stellite type nitrided steels, cemented carbides and borides, flame-hardening, hard chromium-plating and induction-hardening of outer surfaces are among the best known of these recently developed aids to better design and manufacturing practice. The newest addition to this field is a process (the Budd process) for induction-hardening the internal surfaces of cylinders, tubes, pipes, etc., developed in the United States and fully described recently by F. P. Peters and E. F. Cone.\*

This new process, although akin to other methods of producing hard surfaces for wear resistance, is said to be unique in several respects. It is the first commercially successful application of automatic induction-hardening to the inside surfaces of cylindrical parts—an entirely different proposition from external surface hardening; the hardness produced in both steel and cast iron are remarkably high, yet are accompanied by unexpected machinability, physical properties, and microstructure; ultra-high integral surface hardness can be conferred on certain types of industrial surfaces where it is greatly required—Diesel engine cylinder bores, aircraft engine cylinder barrels, wheel-hub interiors, etc.; and the equipment used is a precision machine tool capable of turning out hardened parts on a mass-production scale.

The fundamental purpose of the process is to develop a hard, wear-resisting surface on the inside area of a cylinder without disturbing the toughness, ductility, machinability and dimensions of the main body. A corollary objective has been the production of this condition in ordinary steels

and irons as well as in alloyed material. Actually, the process is now doing these things in large-scale applications, and an automatic control system ensures for each job uniformity of area treated, of depth of hardness penetration, and of hardness along the treated surface.

The selective hardening is accomplished through the concentration of high-power high-frequency electromagnetic current in the surface to be hardened. The high-frequency currents are caused to flow almost entirely in a shallow internal surface zone, and heating is so fast that this zone is raised to hardening temperature almost instantaneously—at least before any significant amount of heat is lost to the main body of the part. The immediate application of a controlled water quench finishes the metallurgical treatment. Close control of results is obtained through accurate automatic control of the power input and frequency of the current, and of the volume pressure and angle of direction of the water quench.

The equipment used comprises a source of high-frequency current (usually a motor-driven inductor alternator), and conducting and transmitting accessories; the hardening machine, with its heating head, quenching fixture, and work-holding device; and a controller system. The heating head, which is the vital part of the machine, is connected with the transformer secondary and comprises a copper tube coiled around a laminated iron core, all on a mandrel. Cooling water is circulated through the tube, since current densities as high as 200,000 amps per sq. in. are reached in some cases. The quenching device operates directly under the heating head and in controllable relation to it. It is attached to the lower section of the machine, rises towards the head during the treating cycle, and is retracted downwards during loading and unloading. For each job a suitable work-holding fixture is provided (usually

\* *Metals and Alloys*, 1941, vol. 13, 6, pp. 713-722.



on the lower part of the machine), and in some cases the part being treated is simultaneously rotated and moved axially with respect to the heating head during the treatment cycle.

In operation the machine is set for any desired heat-treating operation within its range by a convenient adjustment of the controller and its cam-operated switches, and the proper heating head and work-holding fixture. This fixture closes and automatically raises the cylindrical work-piece so that it surrounds the heating head with the latter and the quench fixture in position at the bottom of the bore to be treated. The cylinder is then automatically lowered while the high-frequency current is applied and the quench follows at the correct interval, the treatment thus being "from the bottom up." At a predetermined moment, the quench is automatically shut off, moved away from the heating head and the cylinder is removed. The whole operation—precisely reproducible for an indefinite number of repetitions—may require no more than 3 secs. for the complete cycle.

The process has found important large-scale applications for three types of products, aircraft engine cylinder barrels, cast-iron cylinder liners, and automobile rear hubs, and has also been used on a small scale for hardening the bores of hydraulic cylinders and of slush-pump cylinder liners for pumping colloidal mud in oil-well drilling, as well as for heat-treating oil-well casings to increase joint strengths. In all the applications the convenient machining practice made possible with the new method has been a large factor in its selection. Not only can the outside surface of a unit be machined after hardening the bore, where design permits, but the process also makes possible the practice of composite heat-treatment hardening a bore after the cylinder has been heat-treated throughout and machined to finished dimensions. An outstanding feature of the process is that surfaces hardened to 66-69 Rockwell C can themselves be milled or drilled, using proper machinery technique.

Aircraft engine cylinder barrels of 1% chromium, 0.20% molybdenum steel heat-treated throughout by oil-hardening and tempering to 32 Rockwell C can be induction hardened to 62-64 Rockwell C to a depth of about 0.045 in.-0.050 in. In the hardening of cylinder liners made from a good grade of plain or low alloy cast iron, bore surface hardnesses of 60-66 Rockwell C can be attained to a depth of about 0.070 in. Such hardened surfaces are machinable with special technique and are not in the least brittle. Automobile rear hubs of 0.4% carbon steel hardened by the process give 56-64 Rockwell C hardnesses to a depth of about 0.10 in., and have extremely good wearing characteristics. In general, steel surfaces develop higher hardnesses in treatment than do those of cast iron. Steel parts have been readily hardened to 69 Rockwell C in depths from 0.025 in. to 0.375 in., and certain special cast steels have even been treated to hardnesses up to 78 Rockwell C.

The results obtained from this new process of induction hardening are considered to have their origin in the microstructures resulting from the special treatment. The extremely rapid quenching produces in cast iron and steel a much finer grained martensite than is ordinarily obtained by other methods. In all cases the grain is ultra-refined, ordinarily there remains no vestige of the original grain boundaries. In cast iron the uncombined carbon seems to be recombined at a greatly accelerated rate with attendant increase in hardenability and strength of the area treated. Lamellar structures are also found throughout the hardened positions of originally pearlite irons. Both cast irons and steel show a considerable amount of retained austenite in the hardened zones, in spite of the high hardnesses obtained, and this combination of retained austenite and high hardness may contribute to the wear resistance of the surfaces.

With the process also, distortion is held to a minimum because only a portion of the piece is heated to critical temperatures. Absence of scaling, spalling, grinding cracks, and of relief of hardening stresses during subsequent machining or grinding, and in many cases the obviating of

pre-annealing or pre-normalising treatment, all result from the unusual speed of heating and uniformity of structure.

### United States Bauxite Consumption

DURING 1940 the United States consumed more bauxite than in any other year, according to the U.S. Bureau of Mines. Apparent consumption in the United States in 1940 totalled 958,695 long tons, 22% more than that consumed in the previous peak year of 1939. While the aluminium industry, which produced 26% more metal in 1940 than in 1939, accounted for most of the increased demand for bauxite, artificial abrasive and chemical manufacturers also use more ore. The larger demand in 1940 was met by a 17% increase in home mine shipments and a 21% increase in imports. Home shipments were equivalent to 47% and net imports to 53% of total apparent consumption. About three-fourths of the aluminium industry's supply came from abroad, chiefly from Dutch Guiana.

A special survey showed that in March of this year home bauxite production (55,000 long tons) was running about 50% ahead of the monthly average for 1940, and if that rate continues 1941 output will set a new high record. The survey revealed that home commercial ore reserves were adequate for a short but not for a long emergency, and that production could be multiplied several times within a few months with present equipment at the mines or under construction.

### Oil Temperature Control in Quench Baths

THE increasing use of alloy steels has led to rapid advances in furnace design, including accurate temperature control of the heat. Highly refined and stabilised quenching oils are also now available with suitable viscosity and conductivity characteristics. The importance of maintaining quench baths at pre-determined and specified temperatures, from the start of heat-treatment operations to final shut-downs, is also generally known, but so far little has been done to control the temperature of the quenching oil in use, or to determine the temperature most suited to the particular steel being heat-treated and, in consequence, in some plants inferior work is being produced.

Temperature control of the quenching oil is essential if consistent quality is to be obtained from successive quenches of batches of components, and in this connection interesting information on Prestoild oil coolers is contained in a folder issued by Wild-Barfield Electric Furnaces Ltd. This type of oil cooler, which is thermostatically controlled, takes away the uncertainties which are all too prevalent, and ensures that quenching conditions at the end of the day are the same as those obtained when work commenced.

### Building up Worn Parts by Metal Spraying

THIS process, by which metal parts of machinery may be restored to their original dimensions by a rapid, inexpensive method of metal spraying, is not new; it has been successfully applied commercially in this country since 1922, when the possibilities of producing a hard, coherent mass of metal by spraying were then investigated. If certain precautions be taken, it will be found very successful for the repairing of worn shafts, rollers, cranks, rams and similar parts, thus opening up an enormous field to the engineer in the way of salvage work.

Essentially the process consists in depositing any suitable metal on the worn component by means of a metal-spraying pistol, which uses a specially drawn wire of any desired metal or alloy for its raw material. It will be appreciated, however, that long experience of parts treated was necessary to gain commercial evidence of the value of the process and to develop suitable technique in practice. Very useful information covering the results of experience on this subject over the last decade is given in an interesting brochure issued by Metallisation Ltd., Barclay's Bank Chambers, Dudley, which can be obtained on application.

# Painting Magnesium Alloys

*Several methods are available for the surface protection of magnesium alloy components, but attention is continually being directed to improved methods of protection, and as a result of investigations it is claimed that by proper selection of alloys, surface treatment and painting system, it is possible to obtain satisfactory protection on magnesium alloys for several years, even under relatively severe service conditions. These investigations are discussed.*

**M**AGNESIUM and its alloys are being increasingly applied in industry largely because research and development have supplied satisfactory alloys and practical fabricating techniques. In meeting the requirements of many fields of application, however, protective coatings are necessary if magnesium alloys are to give the best service of which they are capable. The special characteristics of this metal have made painting a problem which has required long and tedious investigation to reach a practical solution. Protective coatings can be produced by chemical or electrochemical treatment of the surface of the metal as well as by the application of paint coatings. The greatest protection is obtained by a combination of both methods.

In investigations of this character, accelerated exposure tests are essential if rapid progress is to be made in evaluating coatings. To be significant, however, the test should give reliable information as to the relative performance of different coatings in service. The salt spray test has been used to some extent, but this test ignores the effect of sunlight on the coatings also other important factors. In an investigation recently recorded\* an accelerated exposure test, comprising cycles of immersion in synthetic sea water combined with atmospheric exposure at an angle of 45° to the vertical facing south, has been found a reliable means of quickly obtaining a measure of the relative durability of various protective coatings on magnesium alloys.

In this investigation casting alloys were used which, in addition to magnesium, had the following nominal composition:—

Alloy	Aluminium %	Zinc %	Manganese %
1	6.0	3.0	0.2
2	10.0	—	0.1
3	6.5	0.7	0.2
4 (a)	6.8	1.2	0.2

(a) With maximum iron and nickel content of 0.005% each.

A rather wide difference has been found in the ability of paint to protect the various alloys. Two panels composed of Alloys 1 and 2 respectively were each given the same surface treatment and paint coating. After nine days in the test the panel of Alloy 1 showed practically no failure, whereas considerable failure of the coating was observed in the Alloy 2 panel. A similar Alloy 2 panel, without a paint coating but having the same surface treatment, was covered with a continuous film of corrosion product after two days in the accelerated test. Without the surface treatment, metal corrosion would have occurred within 24 hours. Some corrosion appeared on the unpainted surface of an Alloy 1 panel after two days' exposure, but to a minor extent. This supports the view that the difference is largely one of alloy composition.

The fact that paint appears to give the best protection on the most resistant alloys is indicated by tests on two panels composed of Alloys 3 and 4 respectively, each of which was given the same surface treatment and paint coating. The panel composed of Alloy 4 was much superior in its resistance to attack by salt water. After four weeks the test on the Alloy 3 panel was discontinued, but the Alloy 4 panel was in good condition after more than 66 weeks in the test.

## Surface Preparation

In common with other metals, adequate preparation of the surface of magnesium alloys is vital to good paint performance. It is especially important in the case of

magnesium alloys which are to be subjected to severe conditions in service, such as contact with salt-laden air in which seaplanes are used.

As in the case of aluminium, films and coatings can be formed on magnesium and its alloys by a variety of chemical or electro-chemical treatments, but magnesium oxide or hydroxide is not so insoluble in water or so impervious as the oxide on aluminium and therefore is not so protective. However, the best of the coating methods now available offer substantial protection to magnesium and its alloys, and are indispensable as surface preparation for the subsequent application of paint.

Probably the best known and most widely used of these methods is commercially termed "chrome pickle," and is applied by immersing the article to be coated in a strong solution of nitric acid and sodium dichromate at room temperature or warmed somewhat. The treatment period varies from about 30 seconds to 2 minutes, depending upon the age of the solution. The article is then removed, rinsed in cold water, followed by immersion in hot water, drained, and dried. Most castings, extrusions, and other articles are given the chrome pickle treatment by the manufacturer, even if they are not intended to be painted.

One of the best protective methods is a two-step process in which the article, after the chrome pickle treatment, is sealed by immersion in a boiling solution of sodium dichromate for 30 minutes. Another effective treatment is the HF-dichromate method described by Schmidt, Gross, and de Long.<sup>1</sup> This is also a two-step process consisting of 5-minute immersion in 15 to 20% hydrofluoric acid at room temperature followed by thorough rinsing in running water. The second step consists of boiling for at least 45 minutes in a 10% solution of sodium or potassium dichromate, followed by rinsing in cold and hot water.

Cleaning of a magnesium article prior to the coating and painting steps is an important element in securing proper protection. Foreign matter or impurities on the surface may be removed by immersion in an acid solution; the chrome pickle solution, which is frequently applied as a routine treatment, may serve this purpose. Grease and oil can be removed with a hot alkaline cleaner, such as one containing trisodium phosphate and sodium carbonate. In the investigations discussed a variety of coatings was investigated as bases for paint. Of these the chrome pickle following by dichromate sealing, and the HF-dichromate coating proved to be about the most effective both from the standpoint of protection and as bases for paint coatings.

## Selection of Priming Coat

The chief requisite for a priming coat of paint, for use on magnesium alloys, is good adhesion; for many applications, however, the primer must also have satisfactory corrosion-inhibitive properties as well as good resistance to moisture penetration; all these factors should therefore be considered in making a selection. Of the pigments listed, zinc chromate alone shows good corrosion-inhibitive properties on magnesium alloys. To be most effective, the zinc chromate should constitute all or essentially all of the pigment content of the primer. Primers pigmented with a mixture of iron oxide and zinc chromate have given definitely poorer protection.

The vehicle must be carefully chosen to obtain a paint highly impervious to moisture; at the same time it must show good adhesion to the metal. Certain of the synthetic

\* Robert I. Wray, *Ind. Eng. Chem.*, Vol. 33, 7, pp. 932-937.

<sup>1</sup> H. Schmidt, W. H. Gross and H. K. de Long, *Trans. Am. Soc. Metals*, preprint, 1940.





# Corrosion of 18-8 Stainless Steel in Sodium Chloride Solutions

Knowledge of the conditions under which 18-8 stainless steels corrode assists in establishing conditions to minimise failure in this and similar steels. The effects of temperature, concentration and pH of aerated sodium chloride solutions on steels of this composition have been studied. The main results are given and a possible explanation of the mechanism of pitting is discussed.

**C**ONSIDERABLE information<sup>1</sup> has been accumulated concerning the factors which influence the corrosion of iron in water and water solutions. From this information a reasonable mechanism for the corrosion of iron and suggestions for corrosion prevention have developed. For stainless steels no such general information is available. The latter steels, which are of increasing commercial importance, corrode ordinarily much less than do iron and steel, yet rapid failures sometimes occur. Reasons for the failures and means to prevent recurrence are not always obvious.

Present day failures are due to relatively deep local penetration or pitting which takes place within, as well as along, the boundary of the grain and is dependent on factors, both internal and external, to the alloy. Under certain conditions 18-8 stainless steel is likely to fail in contact with sodium chloride solutions through formation of deep pits. How such pitting is affected by various conditions external to the metal is outlined by Uhlig and Morrill,<sup>2</sup> who report the results of a study of the effects of temperature, electrolyte concentration, pH, and oxygen on the corrosion of this alloy.

oxygen concentration is extremely low, corrosion of 18-8 falls to nearly zero. This was indicated by corrosion determinations at the boiling point of 4% solutions. As is the case with mild steel, stainless steel appears to corrode more rapidly with increase in temperature and salt concentration and less rapidly with decrease in oxygen concentration. In distilled water, 18-8 was found to lose no weight and did not pit within 24 hours at any temperature up to boiling.

The effect of salt concentration on corrosion of stainless steel of the same composition is shown in Fig. 2. Data for mild steel obtained under identical conditions are shown for comparison. Maximum corrosion for 18-8 steel at 90° C. occurs at 4%, but for mild steel at 2% sodium chloride. Mild steel corrodes appreciably in distilled water compared with zero weight loss for 18-8 stainless steel. At almost saturated salt solution, the rate of corrosion of mild steel by uniform corrosion and rate of corrosion of 18-8 stainless steel, accompanied by pitting, approach one another. It appears that at this concentration of sodium chloride the limited oxygen solubility decreases the corrosion of both mild and stainless steels to corresponding rates.

The effect of pH on corrosion of 18-8 stainless steel (containing 18.58% chromium, 8.48% nickel and 0.05% carbon) at 90° C. in 4% sodium chloride is shown in Fig. 3. Because of the extremes of weight loss, a logarithmic scale

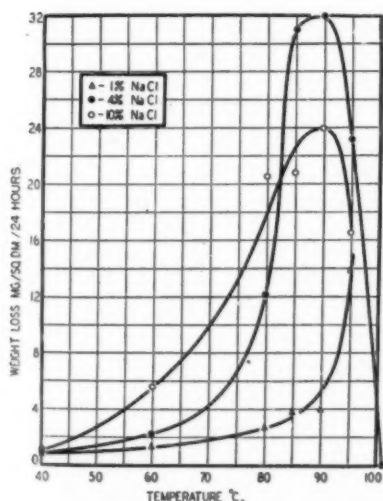


Fig. 1.—Effect of temperature on corrosion of 18-8 in sodium chloride solutions.

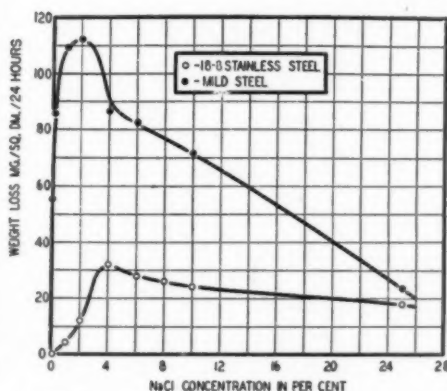


Fig. 2.—Effect of salt concentration on corrosion at 90° C.

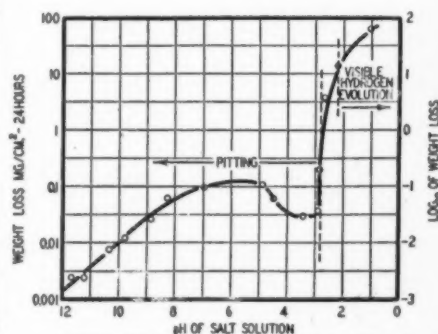


Fig. 3.—Dependence of corrosion of 18-8 on pH of 4% salt solution at 90° C.

The effect of temperature on pitting of 18-8 steel (containing 17.25% chromium, 9.37% nickel, 0.05% carbon) in 1, 4 and 10% solutions of sodium chloride is shown in Fig. 1. Specimens measured  $6.4 \times 2.5 \times 0.35$  cm. and were corroded for 24 hours. Each weight loss is the average of three specimens.

The maximum weight loss for 4 and 10% solutions occurs at approximately 90° C. A maximum undoubtedly exists in like manner for 1% solution, but appears at a temperature nearer the boiling point where oxygen concentration is lower. At the boiling point, or whenever

is used to represent the corrosion data. Each plotted value is the average of three specimens which measured  $5 \times 2.5 \times 0.35$  cm. for corrosion tests in the acid range below pH4, and  $10.2 \times 2.5 \times 0.35$  cm., with some exceptions, for specimens immersed in solutions above pH4. The pH was adjusted either by addition of hydrochloric acid or carbonate-free sodium hydroxide. Because of its sensitive effect on passivity of metals, chloride-ion concentration was kept constant for all tests by omitting an equivalent amount of sodium chloride for any hydrochloric acid added to produce the acid range solutions. Time of immersion was 24 hours, with the exception of determinations at pH values of 4.5, 4.9, 8.3, 9.8 and 10.4. Auxiliary determinations indicated that corrosion weight loss per unit area was sufficiently proportional to time of immersion so that specimens corroded for 48 hours were

<sup>1</sup> G. Henbough, *Chemistry and Industry*, 1933, 195, 228; C. Borgmann, *Ind. Eng. Chem.*, **25**, 816 (1937); J. Friend and J. Brown, *Jour. Iron Steel Inst.*, **83**, 125 (1911); U. Evans and T. Rour, *Proc. Roy. Soc.*, A137, 545 (1932); E. Heyn and O. Bauer, *Mitt. Egl. Materialprüfungsamt, Gross Lichterfelde West*, **26**, 1 (1908).

<sup>2</sup> R. H. Uhlig and M. C. Morrill, *Ind. Eng. Chem.*, **33**, 13, 875 (1941).

comparable in corrosion rate with specimens corroded 24 hours.

Pitting was observed at the edges for all specimens with the exception of those immersed in the very alkaline solutions (pH 11 to 12) and those in fairly acid solutions (pH <2.8). It is probable that longer immersion would produce pits in the very alkaline solutions but not in the fairly acid solutions. In the acid region (pH <2.8) corrosion proceeds by general solution of the surface, with no appearance of definite pits. This process would probably continue unchanged until the metal was all consumed. It was observed under the microscope that specimens immersed in these acid salt solutions were superficially honeycombed, as if many pits had formed.

The number of pits per unit area and the average maximum pit depth determined by a vernier microscope are shown in Fig. 4. Maximum penetration was obtained from the average depth of several pits per specimen and are for 24-hour corrosion periods (white circles). The results indicate that at 90° C. corrosion of 18-8 stainless steel in aerated salt solutions can be diminished if the solutions are made acid to pH 3 or 4, or if made more alkaline than pH 7. Sea-water is alkaline (pH 8); hence in this medium it is expected that, other factors being equal, corrosion of 18-8 steel will be less than in salt solution of equivalent chloride concentration.

The minimum in the corrosion curve for 18-8 between pH 2.9 and 4.5 is the outstanding difference in the curves representing data for the two metals. This differ-

ence is probably attributable to the property of passivity of stainless steels. Iron corrodes by uniform solution of the surface, but 18-8 corrodes by pitting in all but the more acid solutions. The manner in which 18-8 pits is through breakdown of passivity at local areas.<sup>3</sup> The potential difference between passive areas and areas which are active or where passivity has broken down is approximately 0.5

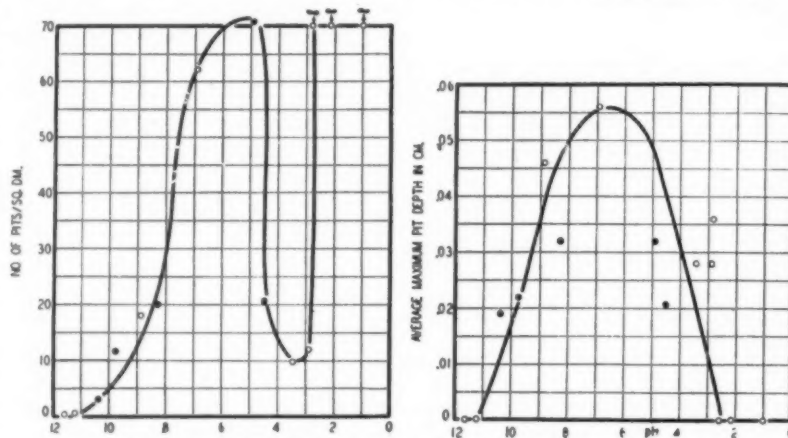


Fig. 4.—Relation between the number of pits and the pit penetration in 18-8 and the pH of 4% sodium chloride at 90° C.

volt. This represents a large potential difference available to motivate corrosion at the active metal areas. An accompanying high current density in the neighbourhood of the corrosion nucleus serves to protect surrounding metal from attack. For this reason, combined with the fact that corrosion products continue to break down passivity, the attack grows deep rather than broad, and pitting results.

Should passivity of the alloy be destroyed, the corrosion proceeds, as in the case of iron, by uniform solution of the surface. There is no longer any tendency for pits to form. Thus, 18-8 steel immersed in hydrochloric acid, which destroys passivity, corrodes not by pitting but by uniform surface solution.

A possible reason for the minimum in the corrosion between pH 2.9 and 4.5 is that passivity of the alloy is partially broken down by increasing acidity of the salt solution. The mechanism is best understood by consideration of corrosion in the alkaline and neutral regions.

The logarithm of corrosion weight loss in the pH 12 to 5 region increases linearly with decrease in pH to a value of 8. The number of observed pits and the average maximum pit depth likewise increase with decrease in pH, as shown in Fig. 4. The increase in number of pits indicates that foci which develop into visible pits are a function of pH. The increase in pit depths, despite increase in the number of pits, indicates that corrosion is more vigorous at lower pH values of the range 12 to 8.

Evidence points to control of corrosion rate by depolarisation brought about by dissolved oxygen in sodium chloride solution. The greater pit depth at pH 8 than at pH 12 indicates that depolarisation is probably more effective at the lower value. The obvious manner in which depolarisation can be more effective, although maintaining oxygen concentration and other variables constant, is to increase the effective cathode area surrounding each pit. The effective cathode area is a function of the electromotive force or driving force of the electrolytic cell producing corrosion; the greater the e. m. f., the greater the cathode area surrounding the anodic pit.

Although the simple electrode reactions involved in the pitting of 18-8 are limited in quantitative interpretation because of irreversible effects, it can be shown that the e. m. f. of the cell increases with decrease in pH. The

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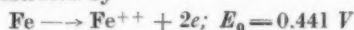
30-35% LIGHTER  
THAN BRASS OR  
ZINC BASE ALLOYS

UTS 22/24 TONS/SQ. INS.  
ELONG. 2/3%

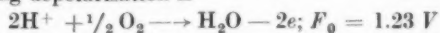
**BROOKSIDE METAL CO. LTD.**  
ASTRA WORKS, HONEYPOT LANE  
STANMORE, MIDDLESEX

<sup>3</sup> H. H. Uhlig, *Trans. Am. Inst. Min. Met. Eng.*, 140, 411 (1940).

anode reaction is the oxidation of alloy metal to its ions, which is illustrated by



The metal salts formed are largely retained within the confines of the pit. Passive alloy in contact with these metal salt solutions becomes active. The cathode reaction involving depolarisation is



and takes place over the surface acting as noble electrode, the area of which is governed by the e. m. f. of the cell. The e. m. f. of the combined electrode reactions is, then,

$$E = 1.67 - \frac{RT}{2F} \ln \frac{(\text{Fe}^{++})}{(\text{H}^+)^2(\text{Po}_2)^{1/2}}$$

where  $R$  is the gas constant,  $T$  the absolute temperature,  $F$  the Faraday, and  $P$  the pressure of oxygen in atmospheres. This equation shows that the higher the hydrogen-ion concentration or lower the pH, the greater is the value of  $E$  and the greater, therefore, are the cathode area and pit growth.

At a pH less than 8, corrosion is less than the extrapolated linear value. This is presumably not due to limitation of oxygen supply at the cathode surface, because it is observed that mild steel (Fig. 2) or other types of stainless steel can corrode in the same solution at several times the rate of 0.1 mg. per sq. cm. per 24 hours at pH 7. There is evidence, however, that in this region passivity begins to break down, and the cathode areas assume less the role of noble electrode and more that of an active metal.

Potential measurements at 25°C. (4) showed that contact of 18-8 with hydrochloric acid of pH 4.09 had no effect, but acid of pH 3.07 destroyed passivity within 10 minutes. Elevated temperatures and increased chloride-ion concentration are known to sensitise further the transformation of passive to active state. Hence at 90°C. and in 4% sodium chloride it is not unexpected that 18-8 will lose passivity in solutions less acid than pH 3.07.

With decrease in passivity of the alloy, there is a decrease in potential difference between the anode areas (pits) and the cathode areas. The result is that effective cathode areas shrink in size. This restricts possible depolarisation by dissolved oxygen diffusing to the metal surface. Because depolarisation by oxygen controls the corrosion rate, this results in a decrease in rate of corrosion.

As passivity is further broken down by increase in acidity of the solution, the effective cathode areas are still further reduced by the diminished driving force of the corrosion cell. Since oxygen diffusion to the cathode surface continues to control the corrosion rate in this region, the corrosion decreases even below the rate at pH 6 or 7. The number of pits and maximum penetration also decrease, the former at pH 5, the latter at pH 7. The decrease in corrosion would be expected to continue with increased acidity were it not for the fact that eventually a pH value is reached at which hydrogen is evolved and oxygen concentration no longer controls the rate.

In the pH 2.9 to 1 region passivity is completely broken down and hydrogen is evolved. The pit depth becomes extremely small, and the number of pits, if such they can be called, increases greatly. As a consequence, the corrosion rate increases sharply and continues a sharp upward trend. Rate of corrosion of 18-8 by hydrogen evolution is of the same order as that of iron.

The sharp distinction in number of pits at pH 2.9 compared with pH 2.8 is clearly apparent, especially under a low-power microscope. At pH 2.9 the pits formed are distinct, and uncorroded metal is observed in between. At pH 2.8 the pits are much smaller but more numerous, and at pH 2.2 the specimen is honeycombed as if a multitude of pits had been induced. With breakdown of passivity it appears that cathode areas have shrunk to microscopic dimensions and pit foci have relatively increased to numbers approaching infinity. The mechanism whereby the alloy corrodes in this state must be similar to the corrosion of iron or steel.

The corrosion products (metal chlorides) within the pit are formed at a pH of approximately 1.5,<sup>3</sup> which lies within this acid range. Hydrogen would ordinarily be expected, therefore, to be a corrosion product for all specimens which show pitting, and weight losses would supposedly include this effect. Except for the region at pH 3 there is no evidence, however, that hydrogen is evolved within the pit. The probable reason is that high anodic current densities at the small pit areas, while corrosion proceeds, bar the proximity of hydrogen or other positive ions to the pit surface. It is conceivable that removal of a pitted specimen from the corroding solution, thereby breaking the electrolytic circuit, or depletion of the corroding solution, thereby diminishing corrosion currents, would result in continued local attack to a slight extent by corrosion products remaining in the pit. This could also occur in the region of diminished passivity at a pH of approximately 3 where the corrosion-cell e. m. f. between surface and pit is much reduced. The slight upturn of the corrosion curve at pH 3.4 is evidence that hydrogen evolution may be taking place, which could occur within the pit, for this is the region of greatest acidity.

The effect of anodic currents mentioned above in limiting hydrogen evolution can be demonstrated by immersing a piece of iron in hydrochloric acid with which it reacts, and making it the anode by connection with an external source of current. With current flow, anodic solution of the iron takes place and hydrogen evolution by local action ceases. The iron electrode in this experiment corresponds to the pit surface in contact with acid corrosion products.

## Chrome Industry in India

MINERALS play an important role in modern methods of warfare. One of the most highly useful minerals from this point of view is chromite, of which India, fortunately, possesses a good supply. The development of stainless steel, containing 30% chromium, has given a great impetus to the use of this metal, and the war has further increased its utility, so that the consumption of chromium is now many times greater than before.

Since the war began attention has been directed to the possibilities of the development in India of a chrome industry on a large scale. A bulletin on mineral chromite, issued by the Geological Survey of India, deals with this question systematically, and discusses the occurrences of this mineral in India, its various uses, the manufacture of chromium, ferro-chrome, chromates, chrome-bricks and various matters connected with grades, prices and the marketing of the ore.

Formerly, Turkey and New Caledonia were the chief sources of supply for chromite, but since 1937 Southern Rhodesia, Russia, South Africa and Cuba have come into prominence; Indian production is less than 5% of the total world production, but the quality is high.

In India, Mysore State is the chief producing centre of chrome-ore, accounting for nearly half the total Indian production, while about one-third comes from the Hindubagh mines near Quetta. Practically the whole of the chrome-ore mined in India is exported, part of it returning in the form of dichromates of potash and soda.

The chief importance of the chromite industry in India has been due to the export trade. But now there appear prospects in other directions. With the development of cheap power, there appears to be a bright future for the smelting of chrome-ore in electrical furnaces to produce ferro-chrome. The production of dichromate is likely to be commenced in the near future by a prominent Indian firm, and it is hoped that this will meet the whole of the Indian demand for this chemical. Considering the progress which is being made in connection with schemes for the manufacture in India of aircraft, car engines and armour-plate, there is little doubt that more and more chromite will be consumed in the manufacture of ferro-chrome in India.



## Tungsten Copper for Electrical Contacts

*Combinations of tungsten and copper offer excellent possibilities for electrical purposes, such as welding electrodes and electrical contacts, due to the physical and chemical properties of the elements.*

ORDINARY metallurgical means do not lend themselves to the production of alloys of copper and tungsten, as the two elements are not soluble in each other and do not alloy in any proportion, so that powder metallurgy has to be resorted to as a means of manufacturing compositions of these metals. Metallic bodies formed of combinations of the two metals are particularly suitable for such electrical purposes as welding electrodes and electrical contacts due to their physical and chemical properties. The physical properties which can be developed in such combinations using various percentages of copper and tungsten have recently been discussed by F. R. Hensel, E. I. Larsen, and E. F. Swazy.\*

Pure tungsten has a specific gravity varying from 19.1 grms. to 19.3 grms. per cubic centimetre, there being considerable variations in the density measurements of powders, rods and wire, particularly if the density is determined on rods which have not been sufficiently worked so as to eliminate all porosity or if tungsten oxide is present. Adopting a value of 19.3 for the density of tungsten powder and a value of 8.92 for copper, and plotting the density of various copper-tungsten combinations as a function of tungsten in volume per cent., a straight line relationship is obtained. The most practical range of densities varies from 12 grms. to 16 grms. per cubic centimetre, and materials with such densities can be produced in comparatively large and complicated shapes and provide a new structural material of high specific gravity. Such a material could be used industrially in the production of fly-wheels in gyroscopes, where a concentrated mass, having a high strength and resistance to creep, is required.

The coefficients of expansion of a series of copper tungsten alloys determined in a hydrogen atmosphere to avoid oxidation of the tungsten show that in heterogeneous alloys composed of a mixture of crystals of the two constituents, the linear thermal coefficient of expansion is a linear function of the volume concentration provided that no solid solutions are formed, and that the alloys are free from internal stress. The theoretical curve of the coefficient of expansion of copper-tungsten mixtures, based on volume per cent., is a straight line. Comparatively low values were obtained lying between that of tungsten of  $4.6 \times 10^{-6}$  and that of copper of  $19.95 \times 10^{-6}$  over a range of temperature from 0° to 600° C. Such low values might be of importance in industrial applications where a material is required which should exhibit the least amount of length or volume changes due to temperature variations.

Experiments show that hardness in copper-tungsten mixtures increases rapidly as the tungsten content increases. The practical hardness desirable in commercial copper-tungsten compounds varies from 60 to 110 Rockwell B, according to the purpose for which the materials are to be used, and the hardness values of well-prepared materials are extremely uniform and consistent. Materials with a hardness of approximately 90 to 100 Rockwell B can be readily machined, provided carbide cutting materials are used, as standard tool steels wear away very rapidly, thereby showing the excellent wear resistance of copper-tungsten compound materials. Using compositions having a lower hardness, copper-tungsten mixtures exhibit good bearing properties and can be used in combination with a softer copper alloy shaft for current carrying bearings.

When the tensile strengths of various copper-tungsten

mixtures are plotted against the percentage of tungsten by weight a curve is obtained having an inflection point between 50 and 60% of tungsten, at which point a rather rapid increase in strength occurs which is accompanied by a rather low ductility. From 55 to 85% tungsten tensile strengths of from 25 to 50 tons per sq. in. are obtained on annealed test specimens. These values are considerably higher than the tensile strengths of either pure tungsten rod or copper of equal section, showing that although true alloying does not occur, a bonding or wetting takes place during the preparation of the materials which enormously increases adhesion.

The compression strength properties of copper-tungsten compounds are considerably higher than the tensile strength. The compression test applied to these materials is a type of transverse test where square test specimens are placed on round supports and a load applied at a point equidistant from the two supports, the ductility of the specimens being measured by the amount of deflection at the stresses causing fracture. The compression strengths of commercial products range from 150,000 lb. to 200,000 lb. per sq. in., depending on the composition, and the compression proportional limit is in most cases 70,000 lb. to 80,000 lb. per sq. in. less than the compression strength. The proportional limit in compression is a linear function of the copper-tungsten mixtures. The behaviour of these mixtures in compression is quite different from pure copper or pure tungsten, as pure copper has no proportional limit, while in pure tungsten the proportional limit and the compression strength are nearly the same.

The electrical conductivity of copper-tungsten mixtures can be more or less calculated by the rule of mixtures, provided that no impurities are present. Recent research work has been successfully directed towards the development of a material having a high electrical conductivity, increased hardness, and tensile strength with some ductility. This material has a tensile strength of 55 tons per sq. in., a compression strength of 210,000 lb. per sq. in., and a Rockwell B hardness ranging from 102 to 108. Special technique is required in the preparation of such material to attain those values, but it is commercially possible to produce copper-tungsten mixtures having a tensile strength of 78 tons per sq. in., a Rockwell B hardness of 110, and an electrical conductivity of 30%. Material with these properties have a specific gravity of 14.5, and form a rare combination of high specific gravity, high tensile strength, and high electrical conductivity, and such materials may find considerable industrial applications, particularly in industries connected with aircraft, automobile, electrical and instrument developments.

Copper-tungsten combinations are particularly suitable for electrical contacts. To a contact metal the combination of a high-conductivity material with one having a high melting point imparts non-welding properties without seriously affecting its capacity to carry large currents. By selecting the proper percentages and ingredients, it is possible to provide the optimum combination of the following properties for specific applications: (1) Resistance to welding; (2) resistance to electrical erosion; (3) low contact resistance; (4) low specific resistance, and (5) resistance to mechanical wear. Copper-tungsten mixtures are also used for facing flash welding dies which have to withstand high pressure and which must have excellent wear resistance. The flash does not stick to the copper-tungsten facings. One of the most outstanding applications of such dies is in the barrel welding industry. The use of a copper-tungsten facing on flash welding dies will increase the life of the dies over that of copper by approximately 10 to 80 times. In addition, the quality and uniformity of the welds are greatly improved. Copper-tungsten mixtures are also used for bushing inserts for current carrying bearings. Solid inserts may be applied for short bearings with small diameters, while for long bearings with large diameters, several longitudinal copper-tungsten inserts are used for each half-bearing shell.

\* *Metals and Alloys*, 1941, vol. 13, 5, pp. 577-585.

